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THE INTRODUCTION OF TITANIUM AT ORENDA ENGINES LIMITED

by

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ORENDA

April 1956

ORENDA ENGINES LIMITED

ENGINEERING and EXPERIMENTAL DEPARTMENT

MALTON ONTARIO

(Member A.V. Roe Canada Limited and Hawker Siddeley Group)

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FOREWORD

The following paper is a summary of metallurgical investigations carried out at Orenda Engines Ltd. during 1954-1955 on titanium alloys. It is primarily related to laboratory investigations initiated at the prototype examination stage or during design data determinations. Reference is made to some metallurgical requirements indicated by service experience of titanium components in an experimental engine.

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INTRODUCTION

In 1950-1951, this company became interested in the potential application of titanium alloys for aircraft engines. The interest at that time was confined to Ti-150A, a Cr-Fe-O alloy for use as blades or discs in the Orenda. Although several blade and disc forgings were actually produced, severe metallurgical problems were encountered (erratic properties, tungsten inclusions, forging cracks, etc). Finally, experimental work ceased in view of the urgency of other problems as the Orenda went into production. Although this experimental work was abortive, it should not be assumed that the experience so gained was completely lost; for instance the value of radiography, an "irregular" method of examining a wrought product, was to be demonstrated later on.

In the Fall of 1953 the existing state of titanium technology and its potential development as an "engineering alloy" was reviewed and subsequently, introduced in our experimental engine programmes.

During 1954 and 1955, some 1325 pounds of sheet and 50,850 pounds of bar, billet and forging stock were processed into engine components according to the following usage:

	Gross Titanium Mill Product Use - 1954, 1955 (In Pounds)				
Application	Fe-Cr-Mo	4Al-4Mn	6Al-4V	5Al-2½Sn	C.P.T.
Rotor Blades	3,743	5,592	160	12	
Stator Blades	3,676	3,342			
Compressor Discs		14,379			
Rotor Spacer Rings		8,752			
Stator Spacer Rings		1,479			
Stator Blade Rings		4,477			
Misc. Rings		1,261		1,373	2,140
Misc. Sheet				410	904
Misc. Bar	3.5	75	30	24	340
TOTALS	7,422.5	39,357	190	1,819	3,384

This is not to say that the passage was uneventful from the metallurgical viewpoint. Severe problems were encountered at certain stages of the normal cycle from sponge to finished part. In some instances, calculated risks had to be assumed in order to enable the engine development to proceed. Fortunately, with one close exception, such risks were justified.

However, in our work we were severely handicapped by a lack of basic information. In the absence of experience one naturally turns to the text books but in the case of titanium few text books existed. The state of titanium technology has been in a constant flux and with a widespread literature of disagreement between authorities, few felt competent to write the definitive reference text. Because of the rapid advances in technology within a few short years, much of the early literature was misleading or inapplicable to our problems. We knew neither <u>what</u> to look for nor <u>how</u> to look for it. Empirical ad hoc testing in a variety of forms became the order of the day and this, as you know, encourages a very vulnerable position in the aircraft industry. It is vulnerable because although you may effect a temporary "fix", the absence of sound engineering principles or understanding to justify a decision permits the possibility of further problems related to unknown minor variations or deviations from the original state.

From practical experience, we gradually began to recognize and appreciate certain phenomena peculiar to those titanium alloys of interest to the engine manufacturer. While the final story will not be written for some years it is felt that an interim record should be made of the metallurgical data and information collected at this Company to date.

The various types of problems encountered may be grouped or classified as follows:-

- 1.0 Instability or embrittlement of properties due to:-
  - 1.1 interstitial impurities
  - 1.2 indigenous alloy composition
  - 1.3 heat treatment
  - 1.4 forging
- 2.0 Forging skin defects
- 3.0 Melting defects
- 4.0 Welding and brazing
  - 4.1 Transformation
  - 4.2 Contamination
  - 4.3 Brazing
- 5.0 Mechanical Properties
- 6.0 Service Experience.

THE INTRODUCTION OF TITANIUM AT ORENDA ENGINES LIMITED

1.0 INSTABILITY OR EMBRITTLEMENT OF PROPERTIES

1.1 Interstitial Impurities

In May, 1954, a U.S. engine manufacturer suffered a flight failure of a titanium blade ring with unusual metallurgical findings in the post mortem examination. It was found that the alloy Ti 150A, a Cr-Fe-O alloy from which the part was made had virtually no residual ductility after a relatively short engine life; it was quite brittle. This event triggered not only a wave of adverse criticism of titanium as an engineering material but also, on the technical side, an energetic laboratory programme to determine the cause and cure. The problem was complex and with no certain explanation; indigeneous composition, heat treatment, service stresses, a rogue melt, various impurities were all possibilities. By that time, this company was well committed on a tight schedule. While we had for other reasons avoided the use of this particular alloy, we were nevertheless using for our blades a near relative - Ti 140A, a Cr-Fe-Mo composition. Practical considerations such as material replacement, time or even facilities and techniques for degassing were severely restrictive and in the latter instance offered no assurance of any cure. Accordingly, it was decided to proceed on the basis that the material was acceptable for limited service, if it displayed a high initial ductility ($\nless$ 25 percent elongation) in the forging stock and prototype blade forgings.

In the following months, a variety of test procedures were tried - simple tensile tests at room and elevated temperature on specimens from blades, residual tensile tests after axial fatigue tests and after discontinued creep tests, impact tests, heat exposure tests for prolonged periods up to 300 degrees C, all yielded negative evidence of an embrittling feature. However, after creep testing at 400 degrees C and 50,000 psi for 100 hours, a change in residual tensile ductility was noted but no values fell below 10 percent which was considered adequate; neither did prior strain affect room temperature fatigue.

It was not until we carried our exposure tests to 425 degrees C (compare figure 1 with figures 2 and 3) and introduced notched bar testing (figure 4) that evidence of instability could be measured. Notched (K_t = 3.9) 20 degrees C rupture tests showed satisfactory life at low stresses, but failed in an abnormally short time at loads greater than 100 percent of the ultimate tensile strength. (1).

By this time, December, 1954, the engine was due to run and accordingly a 50 hour limit was requested. At 35 hours running time, a fourth stage rotor blade failed.

During the autopsy and in the light of our own and extensive work in the U.S., it could be shown that at over 300 ppm hydrogen, the titanium alloy we had was sensitive to hydrogen embrittlement. Conflicting evidence suggesting contributing factors prevented a proof positive conclusion that hydrogen was the sole cause of failure since the following factors were present:-

(a) the presence of a mismachined root radius contributing an excessively sharp notch,

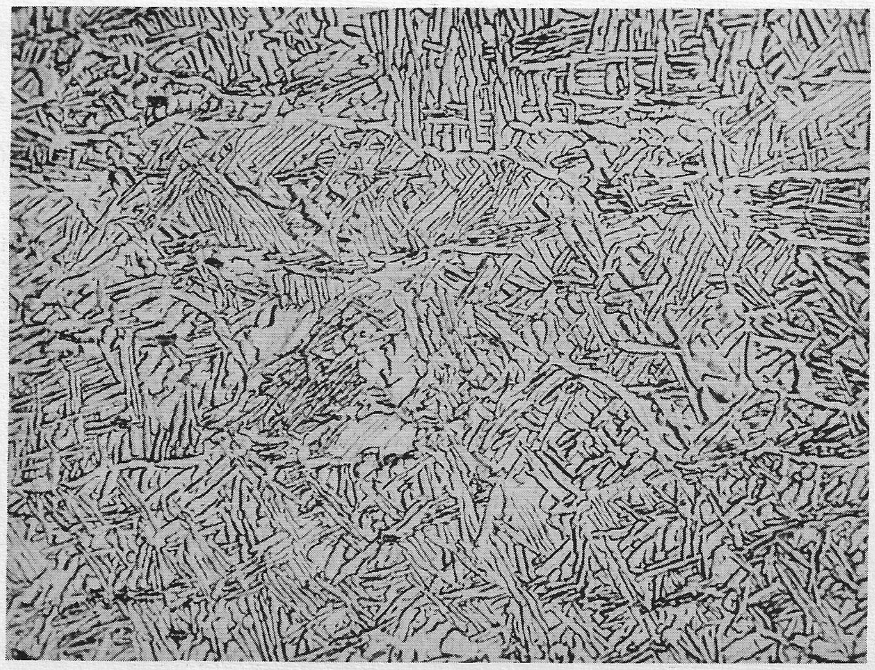
(b) suspected abnormally high service stresses resulting from engine surging,

(c) the absence, metallographically, of subsidiary cracks at the nucleus of the fracture which had been present in all U.S. case histories, and

(d) the relatively low service temperature (less than 200 degrees C) which is considered too low for strain aging phenomena to occur in high hydrogen material.

However, in view of the known possibility of a sudden high tensile load (item b) and the high notch sensitivity of the material in question, particularly in the presence of an excessively sharp notch (item a), the evidence could be largely

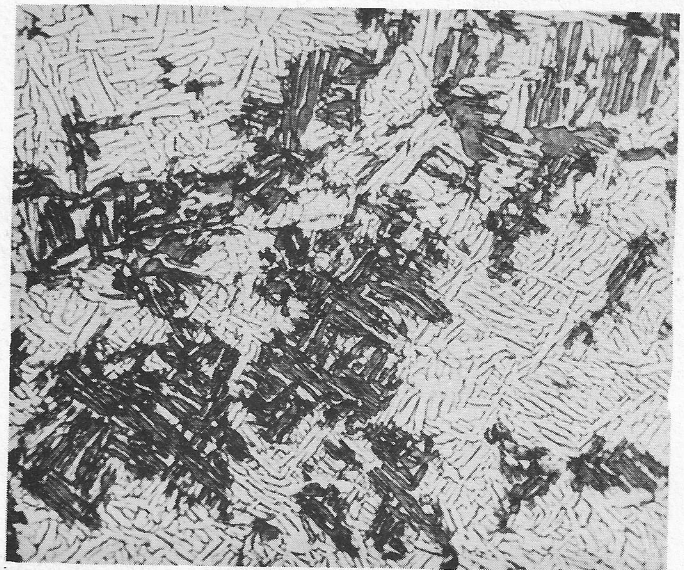
NOTE (1): In a notched tensile test, the breaking load will be greater than that for an unnotched specimen of the same cross section providing adequate ductility exists in the material to permit plastic flow at the notch root; by progressively raising the load above 100 percent of the unnotched strength at periodic intervals, the susceptibility of a material to strain aging may be measured.



X 650 KROLL

Fe-Cr-Mo Alloy (Ti 140A)

As received - hot rolled and annealed 1200°F / 2 hour, A.C.



X640

Fe-Cr-Mo Alloy (Ti 140A)

After 800°F/112 hours

Showing dark etching phase

Fig. 2

X640

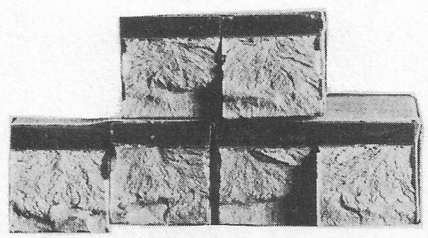
Fe-Cr-Mo Alloy (Ti 140A)

After 900°F/48 hours

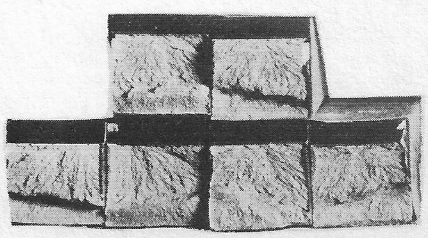
Showing dark etching phase

Fig. 3

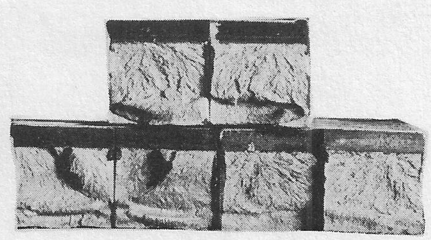
RC 130 B
As Received —



RC 130 B
500°F/300 hours



RC 130 B
700° F/300 hours

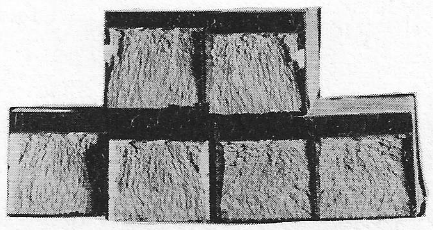


X I 1/3

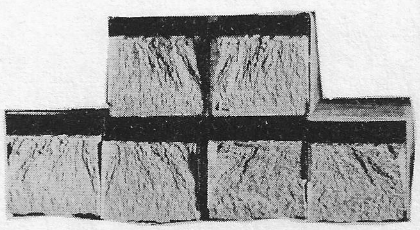
CHARPY IMPACT TESTS, 4 Al – 4 Mn Alloy (RC130B)
values before and after exposure - 12/15 ft.lbs.

Fig. 4

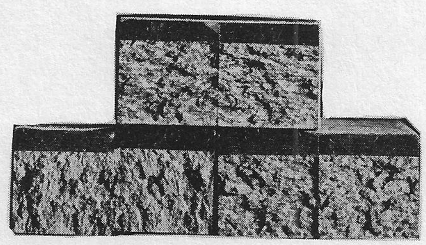
Ti 140 A
As Received



Ti 140 A
500°F/300 hours



Ti 140 A
700°F/300 hours



X 1 1/3

IMPACT TESTS Fe – Cr – Mo Alloy (Ti 140A)
Note coarse fracture after exposure tests 14 ft. lbs., 12 ft. lbs. and 3.5 ft. lbs. respectively.

Fig. 5

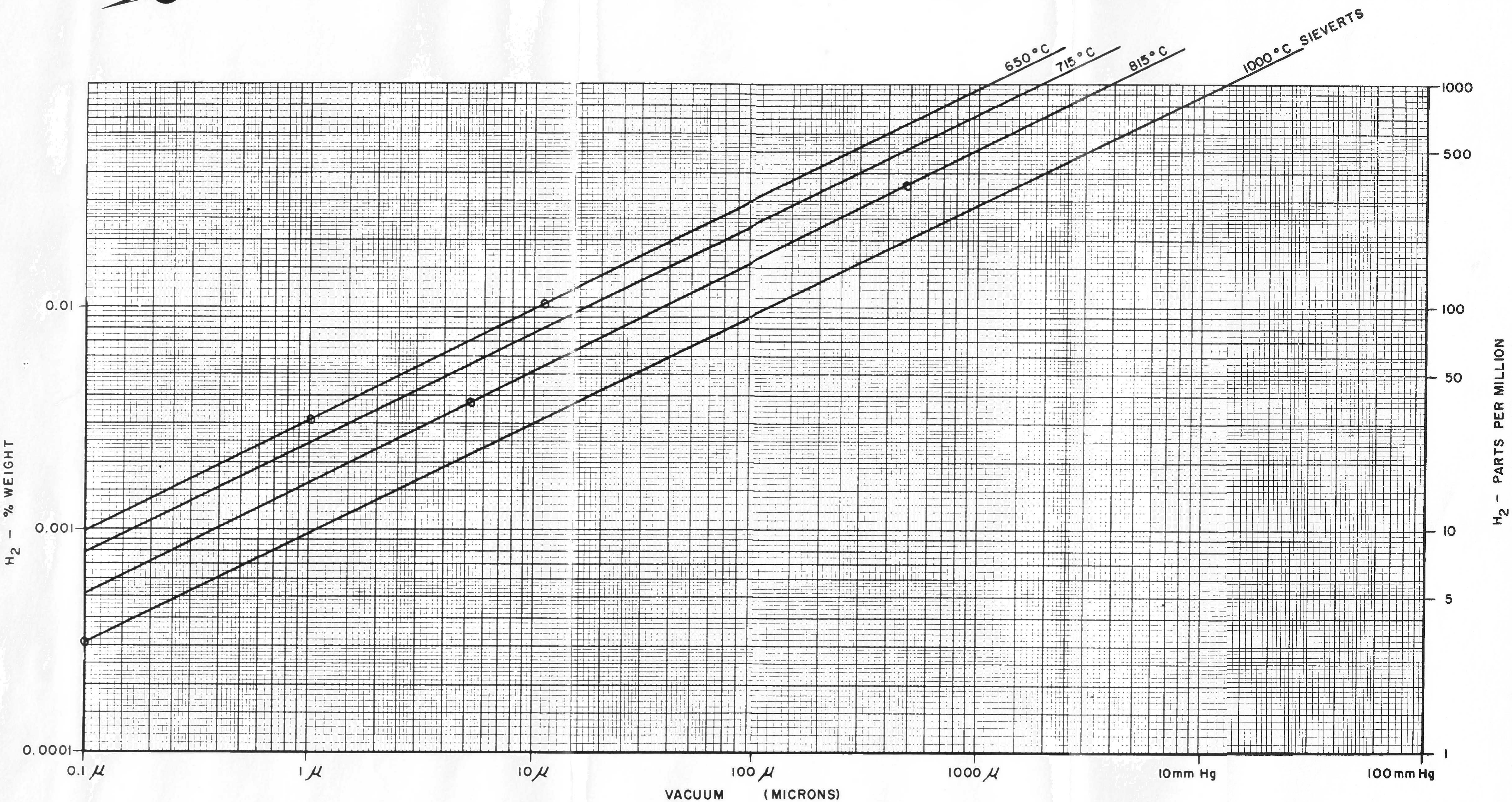

650 °C
715 °C
815 °C
1000 °C SIEVERTS
0.01
0.001
0.0001
H2 - % WEIGHT
1000
500
100
50
10
5
1
H2 - PARTS PER MILLION
0.1 μ
1 μ
10 μ
100 μ
1000 μ
10mm Hg
100mm Hg
VACUUM (MICRONS)

Fig. 6

reconciled and hydrogen was blamed as the major factor in the failure. The remaining blades were degassed according to the hydrogen equilibrium data given in figure 6 before running proceeded.

Although the specific details of any degassing treatment will vary with the size of part processed, it may be of interest to record the following procedure with our compressor blades. A temperature of 1200 degrees F (650 degrees C) for an eight to twelve hour period was selected on the basis of adequate soaking, ruling section size of the blade root, initial hydrogen content (300/340 ppm), dimensional control, batch size (400 to 600 blades) and furnace equipment available. According to figure 6, this treatment would reduce the hydrogen to a level well below the 100 ppm target and 125 ppm maximum desired, providing a vacuum of less than 10 microns was maintained during the time at temperature. The forgings were cooled in vacuum to 500 degrees F (260 degrees C) in order to minimize scaling (hydrogen is removable down to 350 degrees C).

Test specimens run under these conditions with a vacuum of seven microns gave values of 120 ppm (versus 90 possible); with 17 microns, vacuum, 130 ppm (versus 130 possible) was obtained. During the processing of the blade forgings less than one micron was achieved and hydrogen values in the order of 20 ppm were realized. Practical results therefore agreed reasonably well with the laboratory data presented.

The following notched bar test results indicate the improvement realized.

TABLE I

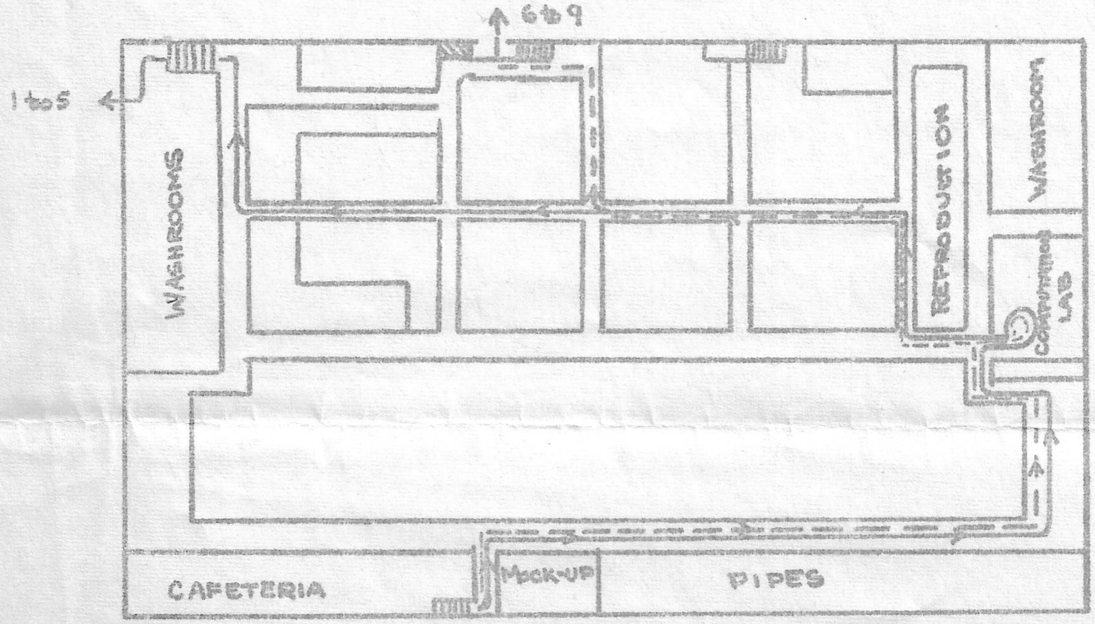
Notch (a) Sensitivity Test Results
Fe-Cr-Mo Alloy (Ti-140A)

Melt No.	H_2 Content (percent)	Stress (b) (psi)	Time
M-1407	0.030	155,000	0.5 hr. F
	0.012	155,000	48 hrs.D
M-1556	0.008	150,000	115 hrs.D
	0.008	150,000	24 hrs.D
M-1352	0.035	157,000	0.2 hr. F
	0.002	157,000	24 hrs.D

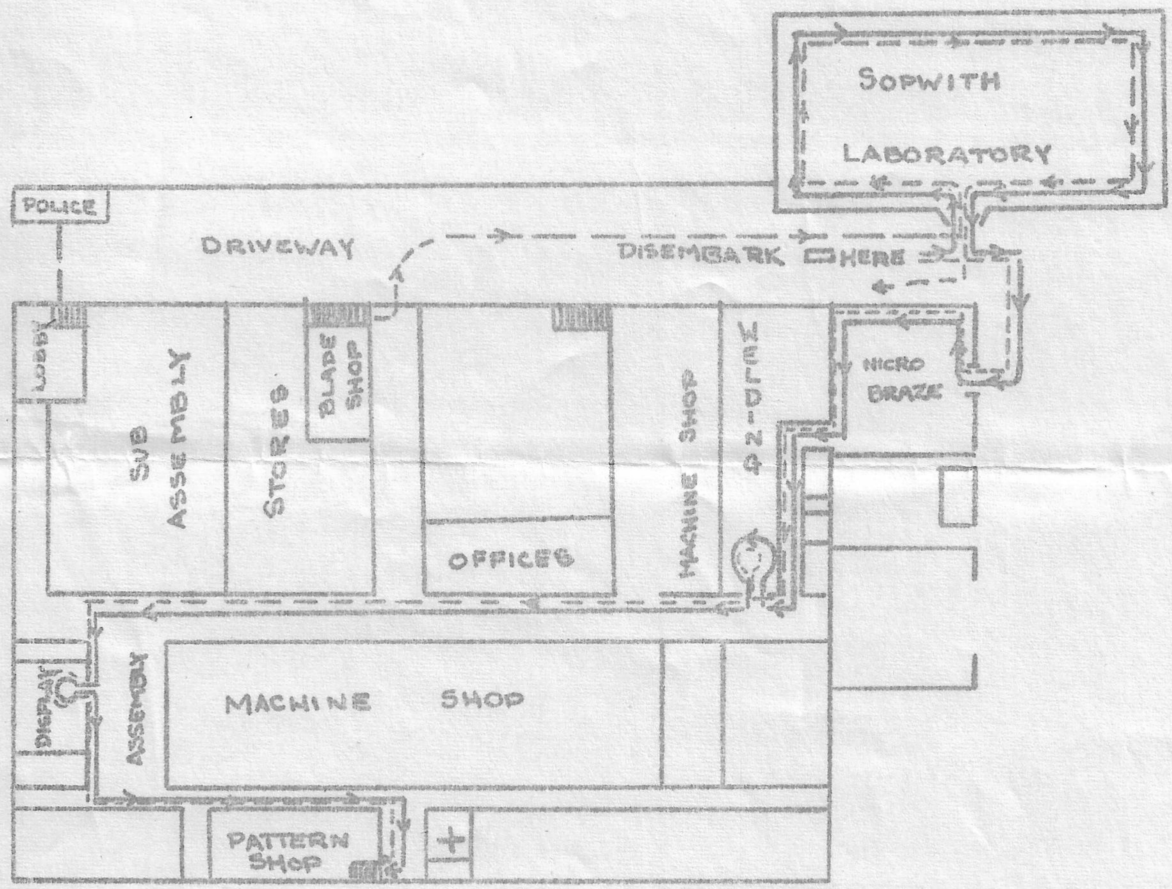
(a) $K_t = 3.9$ F - failed
(b) 110%UTS D - discontinued

K.B.Y.
(6)

ENGINEERING & EXPERIMENTAL BUILDING
ORENDA ENGINES LIMITED



MEZZANINE



GROUND FLOOR

Returning to figures 1 to 5 inclusive, visible microstructural instability is evident in both the microstructure and impact specimen fractures, after exposure for the temperatures and times noted. The dark etching phase in figure 2 could be due to either a complex eutectoid transformation product involving hydrogen, iron or chromium or a simple hydride precipitate. It is interesting to note that repeating the exposure conditions with low hydrogen material, failed to generate the dark etching phase. Figure 3 is included to show a slightly different form of the dark etching phase, similar in appearance to the titanium hydride phase gamma, sometimes found in high hydrogen melts of C.P. titanium. In the case of the Fe-Cr-Mo Alloy (Ti-140A) the impact values fell from 14 to 3.5 foot pounds giving a pronounced woody fracture, in contrast to the 4Al-4Mn Alloy (Rc 130B). In the light of what we know now, microstructural studies and impact tests are relatively insensitive laboratory methods for measuring the effects of hydrogen. The notched rupture test for five to 100 hour periods is acknowledged to be the most sensitive method for measuring strain aging effects due to hydrogen.

As a result of this experience, very careful control during melting and subsequent hot working and de-scaling is maintained to ensure a low level of hydrogen in the finished part. Hydrogen analyses on blades and disc forgings currently being used show a general level of 70 to 90 ppm with some values as low as 40 ppm.

The currently held views on hydrogen, pertinent to metallurgical procedures may be summarized as follows:

(a) In the alpha alloys (Al-Sn alloy, C.P. titanium), hydrogen as the embrittling agent is microscopically visible as titanium hydride and the effect may be measured by tests at high strain rates (impact).

(b) In the alpha-beta alloys (4Al-4Mn, 6Al-4V, Fe-Cr-Mo alloys), hydrogen as an embrittling agent cannot be found microscopically but may be measured by tests at slow strain rates (creep, notched rupture).

(c) Aluminum and possibly tin raise the hydrogen tolerance levels for all alloys. This is believed to be due to the presence of the alpha phase which is increased by the addition of these elements and which has a greater hydrogen solubility.

(d) Vacuum annealing is an effective method for reducing the hydrogen content in all titanium products.

(e) The principal sources of hydrogen contamination have been (a) moisture in the sponge (hydroscopic action of residual salts or acid leaching where employed); (b) atmospheric pick-up during melting and mill processing; and (c) pick-up from acid pickling or sodium hydride de-scaling.

It should be noted that the hydrogen sensitivity or tolerance level may vary the alloy composition. The following tolerance limits have been reported for some alloys of interest to us:-

Designation	Nominal Composition	Hydrogen Tolerance
Ti 150A	Fe-Cr-O_2	70 ppm max.
Ti 140A	Fe-Cr-Mo	180 ppm max.
Ti 155AX	6Al-Fe-Cr Mo	300 ppm max.
C-130AM	4Al-4Mn	160 ppm max.
C-120AV	6Al-4V	260 ppm max.
A-110AT	5Al-2½Sn	265 ppm max.

Aluminum additions seem to increase the hydrogen tolerance, possibly due to its influence on the relative amount of beta present in which hydrogen will dissolve at high temperatures. Similarly, rapid air cooling from the annealing temperature in lieu of slow furnace cooling may increase hydrogen tolerance levels, but other problems may arise with fast cooling (see para. 1.3). The hydrogen tolerance is of interest since it indicates the relative sensitivity to inadvertent deviations in processing and on heavy forgings, the relative problem with high initial forging temperatures or long soaking periods.

Some question might be raised about the individual or interrelationship of <u>all</u> the interstitial elements - carbon oxygen, hydrogen, nitrogen (COHN). In general they are considered undesirable, particularly in alpha-beta alloys where the following limits are usually specified in practice for forgings or mill products.

carbon	-	0.10 percent maximum
oxygen	-	0.25 percent maximum
hydrogen	-	0.0125 percent maximum
nitrogen	-	0.05 percent maximum

They do strengthen the alpha phase (and hence find value particularly in all-alpha materials - C.P., Al-Sn in increasing general strength levels) but have an adverse effect on ductility and impact. They do not promote heat treatability. Some authorities consider a total COHN limit of 0.030 percent maximum as desirable, particularly in alloys designated for welding.

As measured by ductility in a simple tensile, little oxygen influence is noted in annealed 4Al-4Mn alloy until contents of about 0.40 percent are reached. In the solution treated and aged condition, ductility may be more sensitive to oxygen contents at the lower levels.

On the basis of some isolated test results there is reason to believe that notch sensitivity is encountered in the 4Al-4Mn Alloy with oxygen contents of the order of 0.30/0.35 percent (others have reported a correlation in 6Al-4V alloy between poor notch rupture results and oxygen at 0.25 percent). The following table shows the effect of oxygen as measured by the high strain rate, impact test:-

TABLE II

The Effect of Oxygen
On the Impact Strength of 4Al-4Mn Alloy. (2)

	Oxygen Content					
Temperature (degrees F)	0 Percent	0.1 Percent	0.2 Percent	0.3 Percent	0.4 Percent	0.5 Percent
0	13*	12*	10*	7*	4*	3*
100	19	14	12	8	5	3.5
200	30	23	15	12	8	5
300	48	46	25	17	13	7
400	60	54	44	23	18	10
500	67	58	52	25	21	12

* ft-lbs.

NOTE (2): D.R. Luster, B.L. Shakely "How Titanium Alloys Behave at High Temperatures", Iron Age 175 (12) 1955.

The oxygen limit of 0.25 percent maximum is a matter open for considerable discussion. The analytical techniques available are very time-consuming and expensive for routine control purposes and lack close accuracy. A check limit of 0.05 percent is not uncommon. Further, a controlled amount of oxygen may even be desirable for low temperature properties. The effect of oxygen on the recrystallization temperature, a matter of some importance for annealing forgings, is not clearly defined. There is no question that the influence of oxygen and its interrlation with the other interstitials on properties and the control of oxygen in processing including analytical techniques are problems in titanium technology which require further study.

The hydrogen limit quoted is easily attained on barstock without vacuum treatment but on sheet, where a high surface/volume ratio exists and hence more likelihood for pick-up, a higher figure approaching 0.018 percent is more realistic; this higher figure is of course related to the hydrogen tolerance level or sensitivity of the alloy in question.

1.2 Instability Due to Indigenous Alloy Composition

In the course of the work reported above, an unusual microstructure had been obtained after heat exposure tests at 425 degrees C for 300 hours (figure 2). Repeating the tests on degassed material with hydrogen, less than 50 ppm failed to give a similar microstructure but mottling of the matrix suggested the presence of a fine unresolved phase.

Creep tests confirmed the view that an inherent or indigeneous instability existed in the basic alloy composition (Fe-Cr-Mo) of Ti 140A even with low hydrogen. For example, the following data on residual tensile elongation (originally 26 percent) was derived on hot rolled and annealed low hydrogen Ti 140A barstock after creep testing at 45,000 psi for the time and temperature noted.

TABLE III

Residual Ductility of Fe-Cr-Mo Alloy After Creep Tests at 45,000 psi for the Time and Temperature Noted

Time (hours)	550 degrees F (percent)	650 degrees F (percent)	750 degrees F (percent)
16	26.8	26	23.4
50	21.4	12	11
100	27.0		10
250			5.6
500	22		8.8

It can be seen that there is a progressive embrittling effect with time and temperature. Ti 155AX (an Al-Fe-Cr-Mo Alloy from the same family) showed no evidence of a drop in room temperature tensile elongation under similar test conditions until the test temperature was raised to 750 degrees F and time periods approaching 1000 hours were employed. Under these conditions the ductility fell from an original figure of 15 percent to a value of six percent. However, no laboratory work was carried out by ourselves to discover the influence of annealing treatment which may be a contributing factor (data in Table III is for the mill annealed condition i.e. 1200 degrees F/24 hours air cooled).

Phase diagrams published by the research institutions indicated the formation of eutectoid products between titanium and Fe, Co, Cr, Ni, Cu, Si, H_2 and Mn additives. Of these eutectoid forming elements, only the Mn eutectoid has not been observed in practice and is considered to have such a sluggish rate of reaction that it is not a practical consideration. Beta eutectoid reactions are accelerated by high interstitial content.

This introduces the phase or equilibrium diagrams which for the alloys of technological interest to engine manufactures are shown as figures 7 to 10 inclusive (3) and in the simplified diagram of figure 11 (4). These are the basic working data of the metallurgist who should study them carefully for they form the basis of alloy development and heat treatment studies. The following table provides a classification of the alloying function of various elements.

TABLE IV

Classification of Alloying Elements

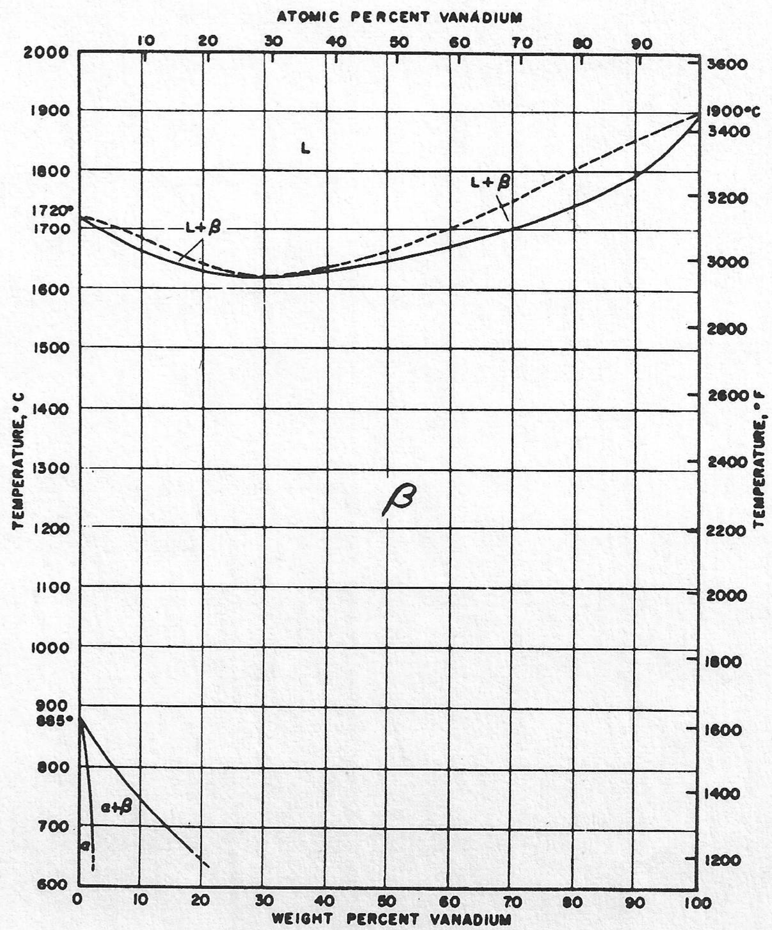
Alpha Stabilizing		Beta Stabilizing	
Interstitial	Substitutional	Interstitial	Substitutional
C O_2 N_2	Al Sn	H_2	Ag, Cb, Co, Cr Cu, Fe, Mn, Mo Ni, Sn, Ta, V, W, Zr.

The beta eutectoid transformation for iron and chromium are indicated in figure 10 as 600 degrees C and 675 degrees C. This would seem to be above any service temperature range. However, the temperatures noted are for equilibrium or idealized conditions and in practice strain aging or interstitials (hydrogen) accelerates the transformation kinetics (see figures 2 and 3).

It may be appreciated that metallurgically, the elements aluminum, vanadium, molybdenum and tin are favoured as alloy additions for the absence of eutectoid forming tendencies;

NOTE (3): M. Hansen, D. McPherson and W. Rostoker, "Constitution of Titanium Alloy Systems" W.A.D.C. Report 53-41, Feb, 1953.

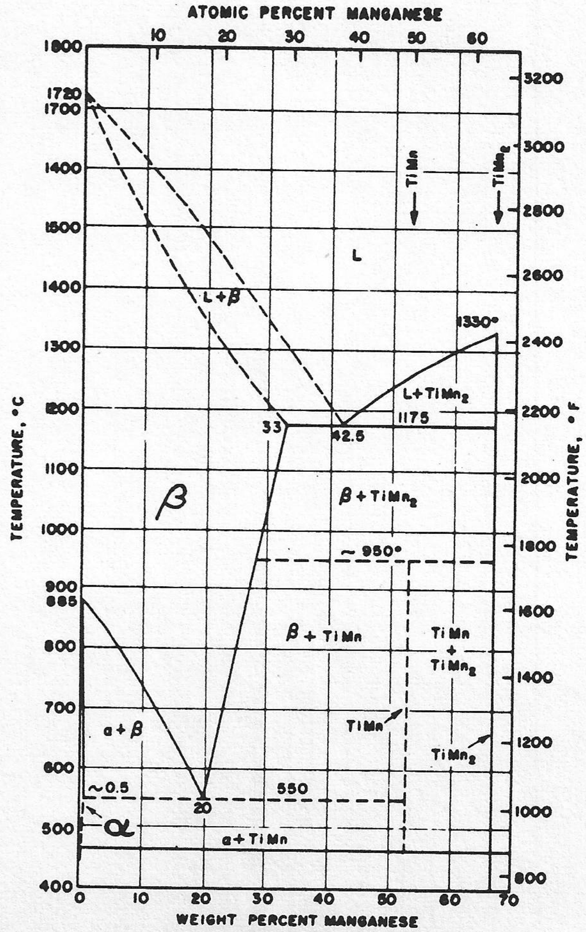
NOTE (4): P. Frost "Metallurgical Considerations in Titanium Bolt Manufacture" N.A.S.C. Titanium Symposium, Oct. 1954.



Ti – V – DIAGRAM

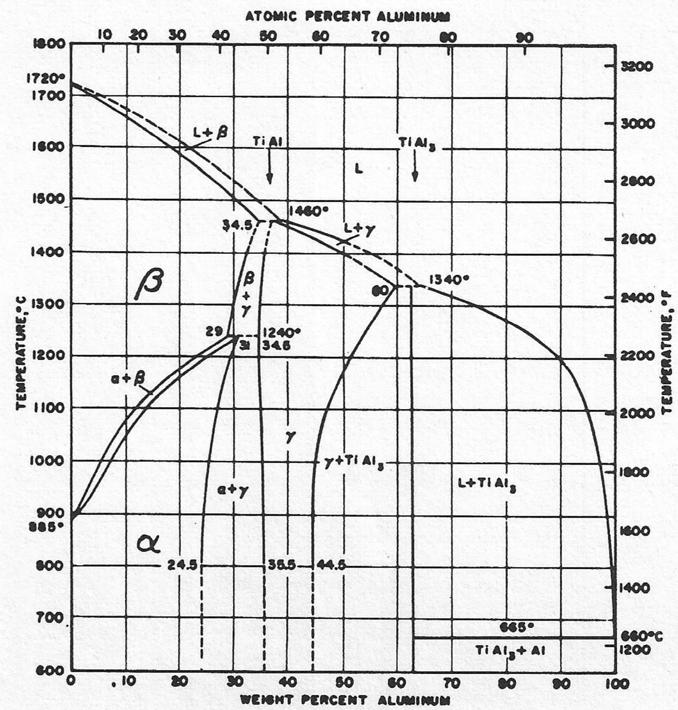
Complete solubility in beta, limited solubility in alpha

Fig. 7



Ti–Mn DIAGRAM

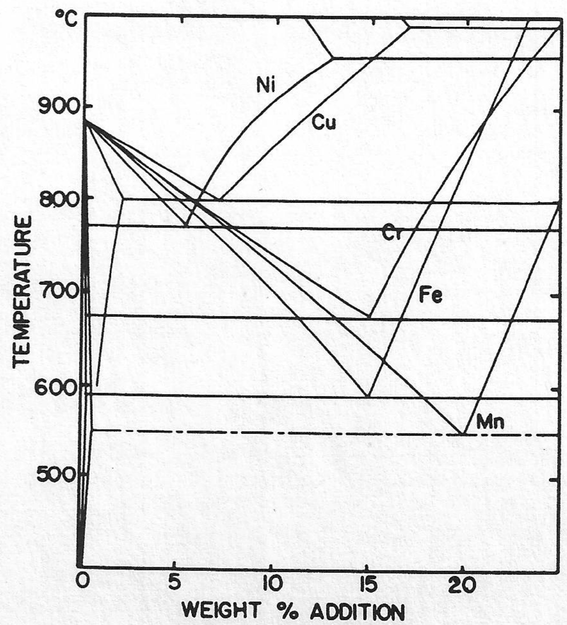
Partial solubility in beta, very limited solubility in alpha.



Ti – Al DIAGRAM

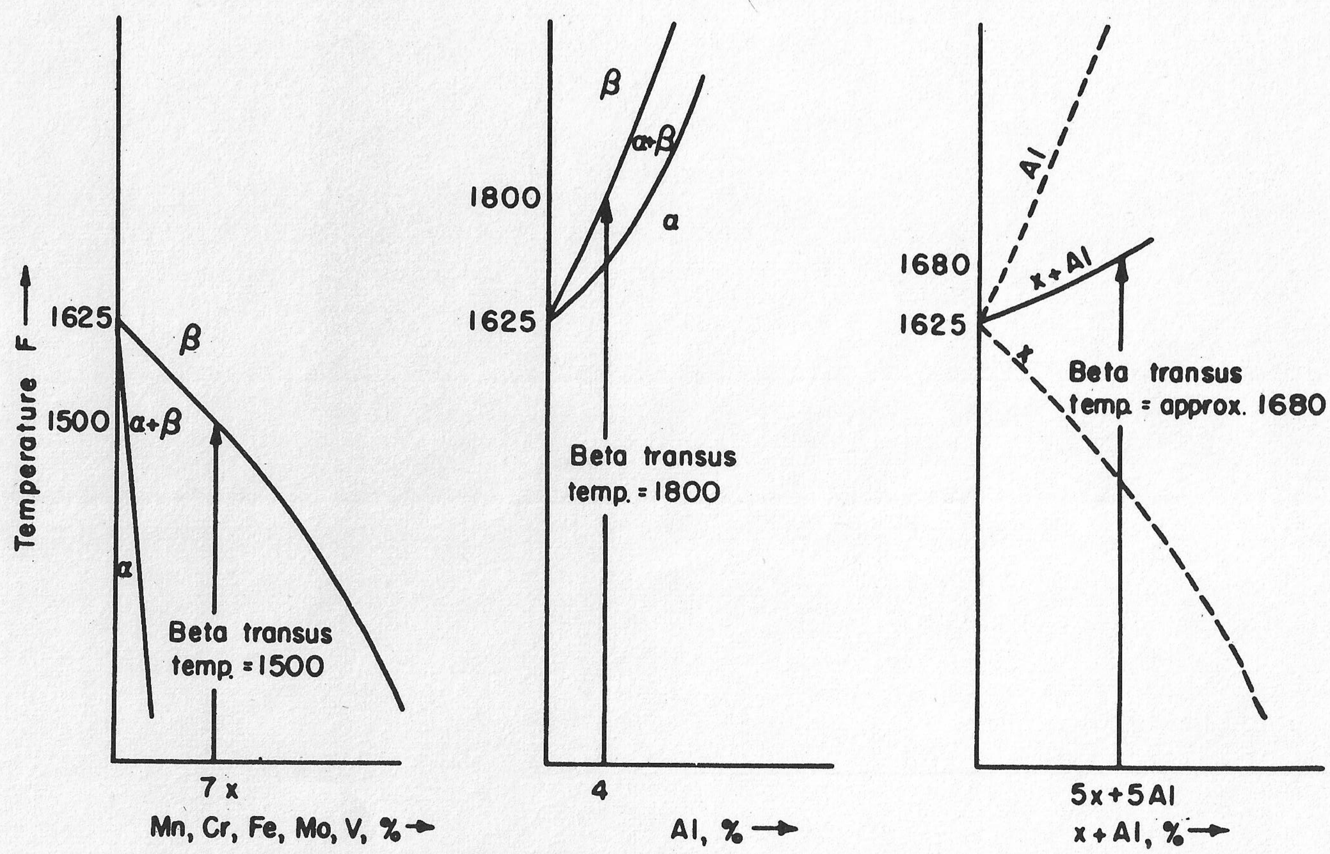
Limited solubility in beta, limited but higher solubility in alpha.

Fig. 9



Titanium Rich End of Binary Diagrams With Copper, Nickel, Chromium, Iron and Maganese

Fig. 10



EFFECT OF TRANSITION ELEMENTS AND ALUMINUM ON BETA TRANSUS TEMPERATURE OF TITANIUM. IN FORMING $\alpha-\beta$ ALLOYS THE FINISHING TEMPERATURE SHOULD NOT EXCEED THE BETA TRANSUS manganese, as already noted, is acceptable due to the extreme sluggish nature of the reaction. Complexing the alloy additions lessens the tendency for composition instability and for other reasons, such as heat treatability, may be desirable. However, the use of such alloys at temperatures over 200 degrees C is not recommended and hence they are of limited interest to the engine designer.

1.3 Instability due to Heat Treatment

It will be appreciated that in our work of isolating the various causes of instability, we travelled many blind streets and that our analyses did not follow the orderly sequence of events presented here. For instance, in our work on composition instability, we were distressed to find that not only the Fe-Cr-Mo alloy but also the Al-Mn alloy showed evidence of changes in mechanical properties after creep strain tests; yet we were assured that manganese was so sluggish an eutectoid former that this reaction had never been observed in practice. At that time we had not heard of omega embrittlement, the intermediatory phase in the transformation of beta to the stable room temperature alpha form. I dare say that if we had known about omega then, it would only have served to confuse us.

The selective reader still may find references to support the fact that the 4Al-4Mn alloy should be annealed at 1200 to 1300 degrees F "and air cooled to provide a stable ductile structure". Moreover, analysis of room temperature tensile test results, on which most quality control work is based, may even suggest that the high side of range (1300 degrees F or even 1400 degrees F) improves the ductility of this alloy significantly. However, careful metallography and consideration of the phase diagrams will show that the aforementioned annealing practice, while perhaps useful for processing mill productions should not be used for forged components.

The problem first came to light when one vendor found it imperative to straighten or size blade forgings after annealing, in order to retain dimensional control. The straightening operation involved air cooling from a temperature of 1300 degrees F.

Figure 12 illustrates metallographically that air cooling results in untransformed beta. The higher tbe temperature (with air cooling) above say 1100 degrees F, the greater the amount of untransformed beta and hence, instability in service.

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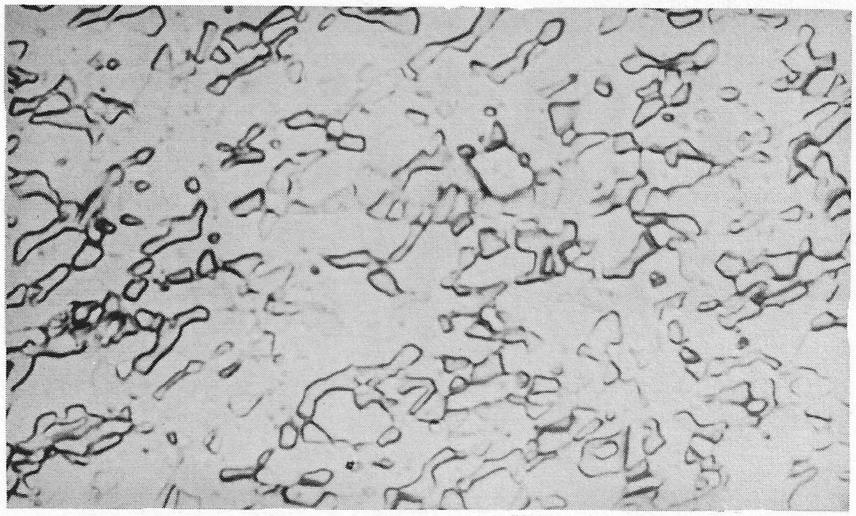
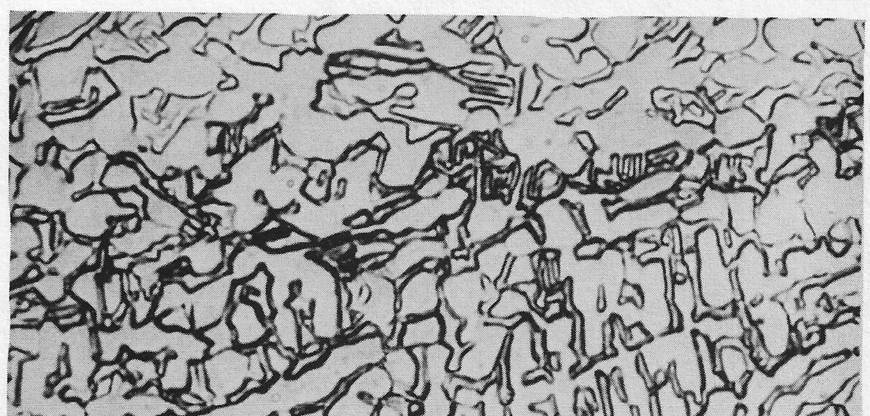


X1300 MOD KROLL ETCH

4 Al – 4 Mn Alloy – Vendor C

After annealing and straightening at 1300°F, A.C.–
Note absence of lamellae of beta and secondary
alpha.



X1300 MOD KROLL ETCH

4 Al – 4 Mn Alloy – Vendor C

After straightening and annealing at 1300°F, F.C. to 1000°F.

The variation and excessive creep rate as well as pronounced changes in residual tensile properties obtained with air cooled material are shown in figure 13.

Illustrated in figure 15 is a form of embrittlement related to heat treatment which because it cannot be seen metallographically, eluded recognition for some time. It is the transition phase, omega. On taking the air cooled 4Al-4Mn material discussed previously and aging it at 900 degrees F for eight hours, satisfactory creep resistance and stable tensile properties were obtained with no visible change in microstructure.

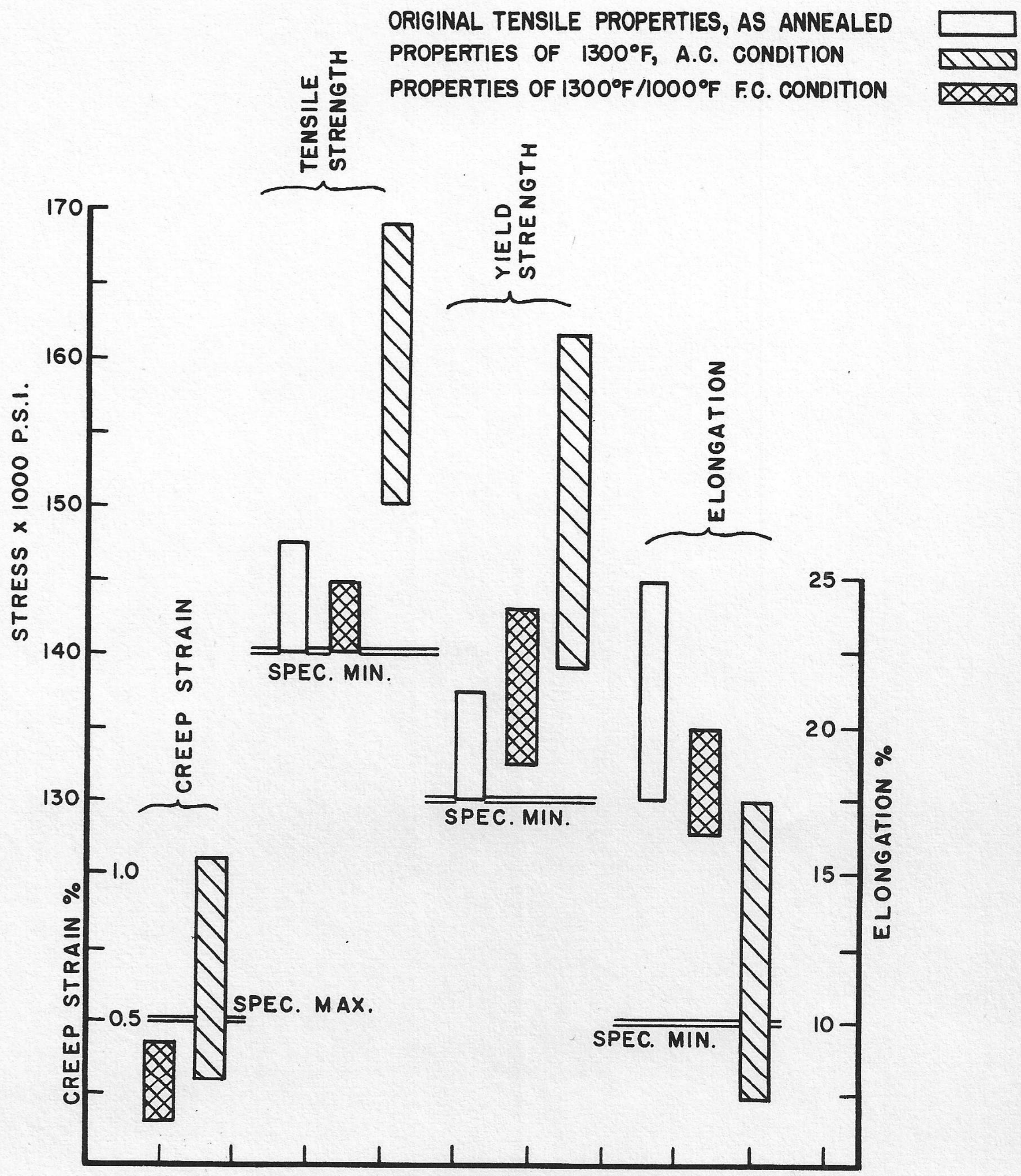
It can be seen therefore, that stability of properties in the blades could be obtained by either of two treatments i.e. furnace cooling following sizing or reheating to 900 degrees F. One treatment is related to the direct transformation or rejection of alpha from beta; the other to the submicroscopic transition of beta to omega to alpha.

This data also defines the stress relieving treatment, which for an alpha beta alloy should not exceed 1150 degrees F; above this temperature, increasingly greater quantities of beta will form which may remain untransformed on cooling, leading to instability of properties.

There are some potential advantages in employing high annealing temperatures (1450/1500 degrees F) in conjunction with furnace cooling to 1100 degrees F. Such a treatment will ensure complete partition of the various elements present (both interstitial and substitutional) between the respective phases and secondly, it will ensure complete recrystallization. This latter factor is not to be overlooked where a combination of high interstitial content (particularly oxygen), low finishing temperature and high reduction during hot working gives you high recrystallization temperatures. The latter conditions are very likely to be present in sheet where the high surface area to volume ratio results in high interstitial content from processing and may be present also in blade or disc forgings where the forging stock has a high initial oxygen content. Balanced against the use of high annealing temperatures for finished forgings are the practical considerations of dimensional control, scaling, sub-scale diffusion (alpha skin effects) and an increase in hydrogen level which are discussed in the following pages. In general, such treatments are more applicable to mill products than forgings. Annealing by furnace cooling from 1400/1500 degrees F will result in lower hardness and tensile strength (say 10,000 psi) but higher ductility than an annealing temperature at the customary 1300 degrees F.

4 AL-4 MN ALLOY
RESIDUAL TENSILE PROPERTIES

AFTER CREEP STORAGE AT 400°C, 50,000 P.S.I. FOR 100 HRS.



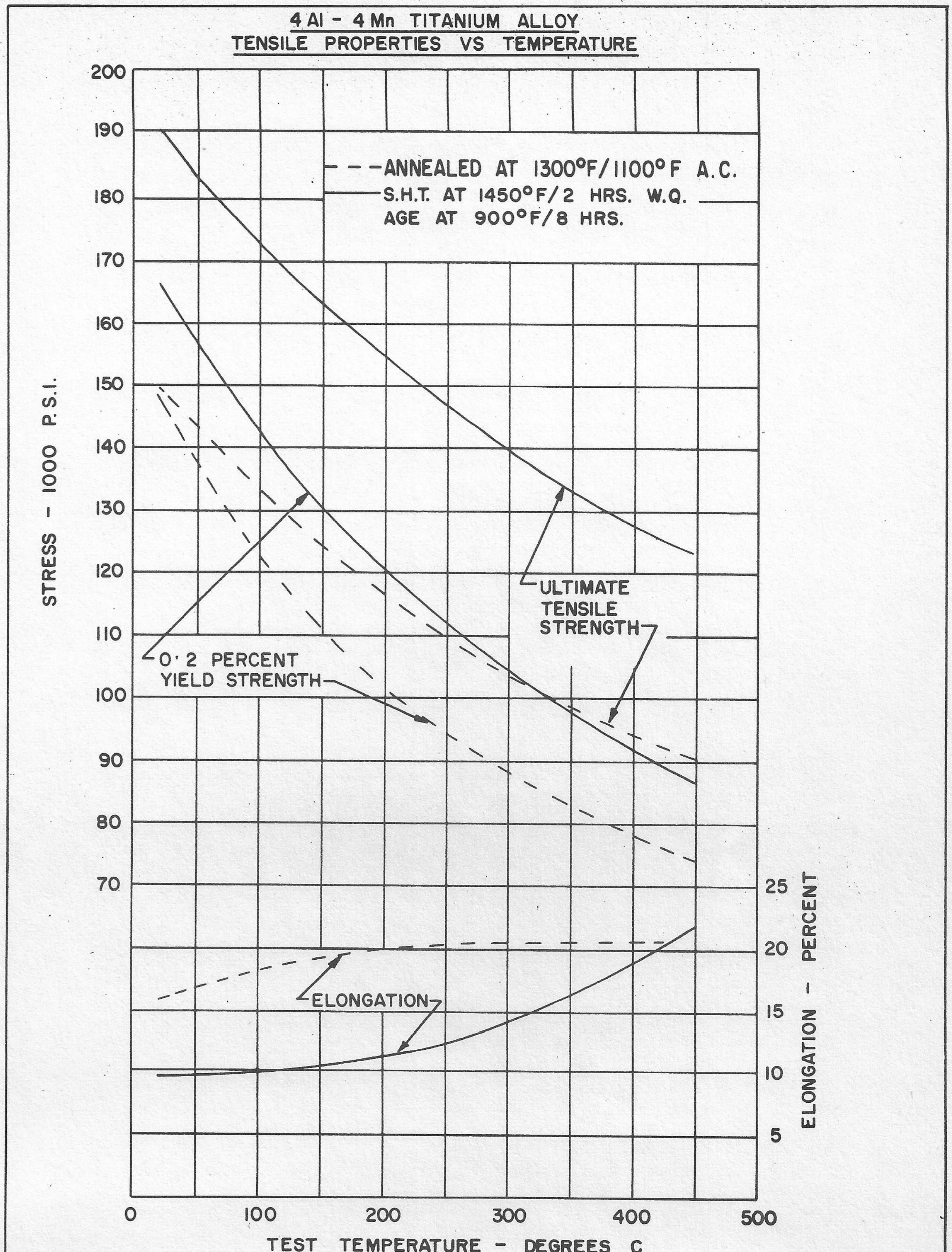
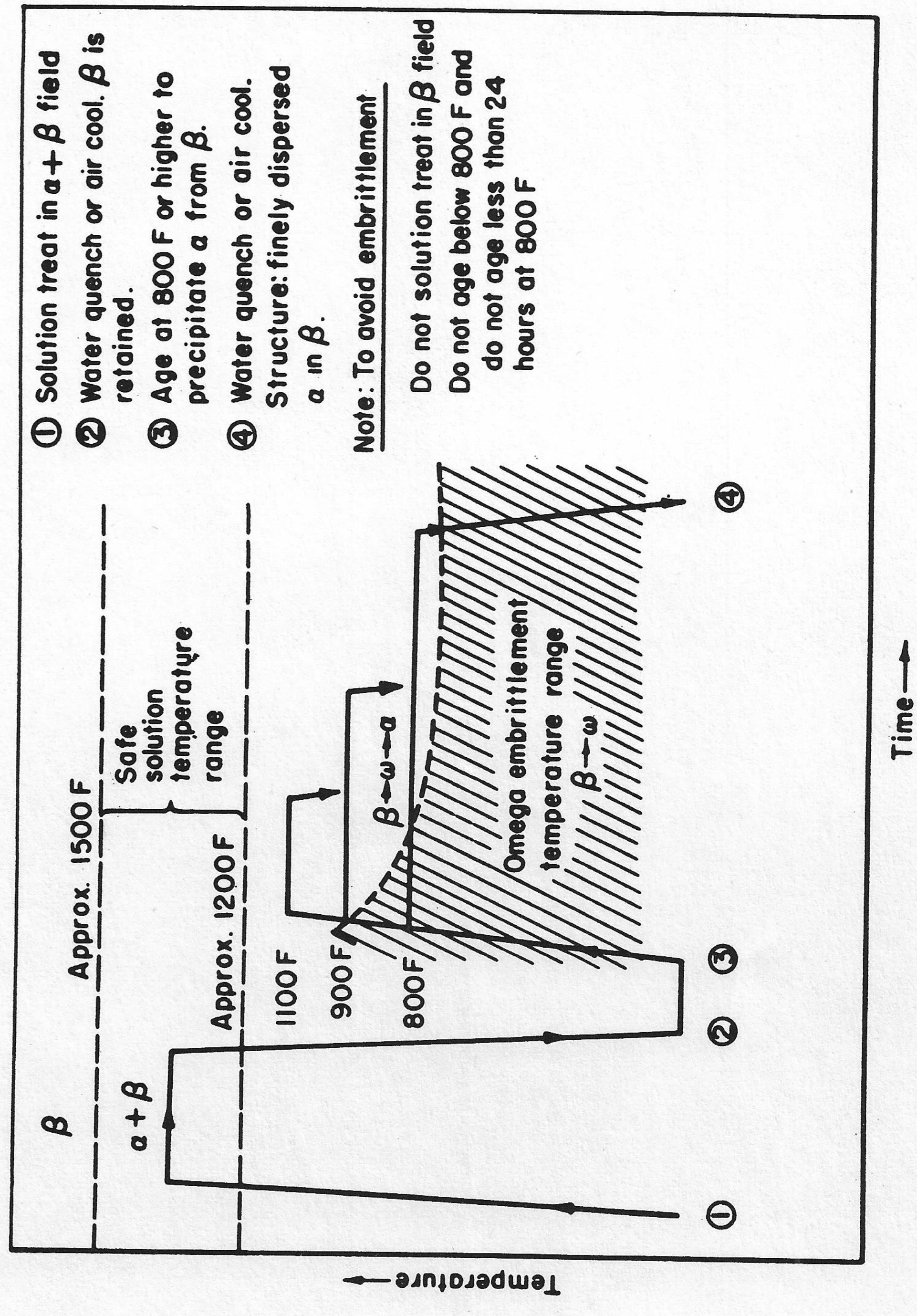

4 Al - 4 Mn TITANIUM ALLOY
TENSILE PROPERTIES VS TEMPERATURE
STRESS - 1000 P.S.I.
200
190
180
170
160
150
140
130
120
110
100
90
80
70
- - -ANNEALED AT 1300°F/1100°F A.C.
S.H.T. AT 1450°F/2 HRS. W.Q.
AGE AT 900°F/8 HRS.
ULTIMATE
TENSILE
STRENGTH
0·2 PERCENT
YIELD STRENGTH
ELONGATION
ELONGATION - PERCENT
25
20
15
10
5
0
100
200
300
400
500
TEST TEMPERATURE - DEGREES C

Fig. 14



RECOMMENDED HEAT TREATING PROCEDURE FOR α–β TITANIUM ALLOYS

Fig. 15

To make a virtue out of a sin, the presence of untransformed or retained beta, while deleterious from the stability viewpoint, in the annealed condition has an important advantage in that it permits a hardening type of heat treatment for alpha-beta titanium alloys, i.e. solution treatment and age. By suitable adjustment in the aging treatment (temperature and time, see figure 15) the unstable transition product omega is avoided and an attractive gain in properties may be realized (figures 14 and 16 refer) at the lower end of the temperature range. This is a nucleation and growth reaction. Unfortunately, hardening heat treatments of this type for the commercially available alloys suffer three demerits in the eyes of the engine manufacturer, namely:-

(a) While a gain in tensile properties is indicated, the hardening treatment offers no particular advantage in creep characteristics above approximately 315 degrees C. This property is important in disc applications.

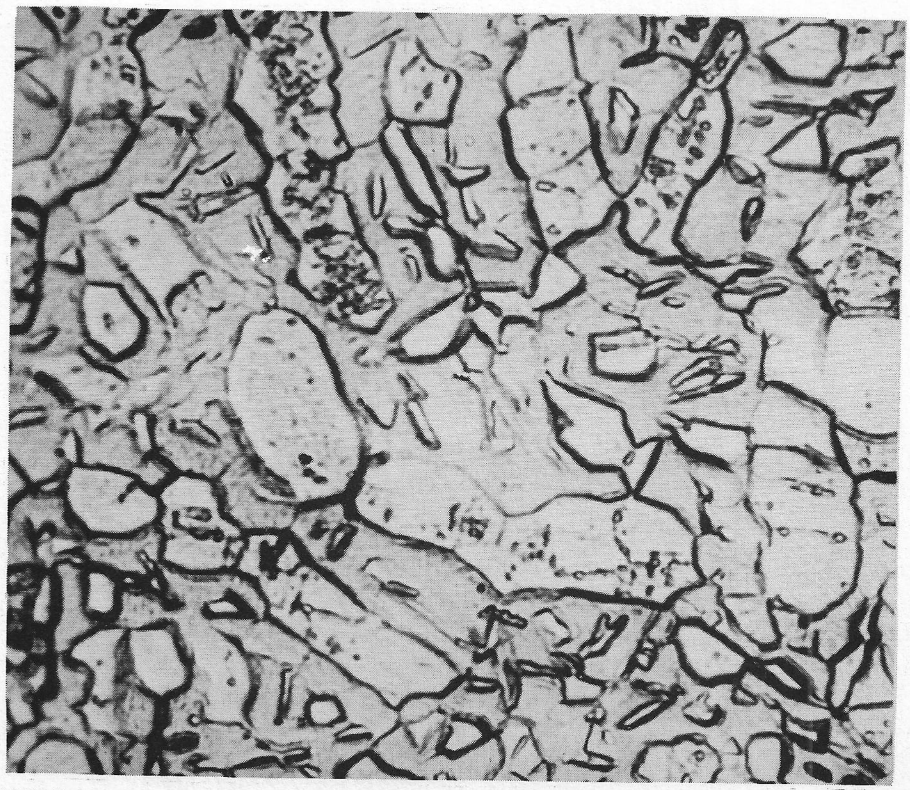
(b) The tensile ductility at room temperature on this series of tests on actual forgings failed to meet the 10 percent minimum, considered to be necessary for a component of this nature. It is not known whether the properties are repetitive and predictable, particularly in terms of an acceptable level of ductility, since isolated low values of five percent and even lower were obtained in other work. It has been suggested that the 6Al-4V alloy is much superior in this regard.

(c) The dimensional control problem on jewelry items such as compressor blades poses a severe problem according to the blade geometry and this problem would be aggravated by quenching and aging.

All of the presently proposed commercial heat-treatments are based on quenching from the alpha-beta field in order to obtain adequate ductility (due to the presence of primary or continuous alpha matrix) with the higher strengths associated with the transformation of the beta phase. The factors governing aging can be clearly seen in figure 15.

The heat treatment fanatics will point out the possibility of quenching from the beta field and hence figures 17 [5] and 18 have been included to show the trends to be expected from alloying elements, cooling rate, intersitial content, aging time and isothermal effects when using such hardening temperatures. However, for the reasons noted above and due to the grain size effect, the best combination of strength and ductility is associated with quenching from the upper end of the alpha-beta region.

NOTE (5): L.D. Jaffe, "Choosing a Titanium Alloy" Metal Progress, March 1955, page 105.



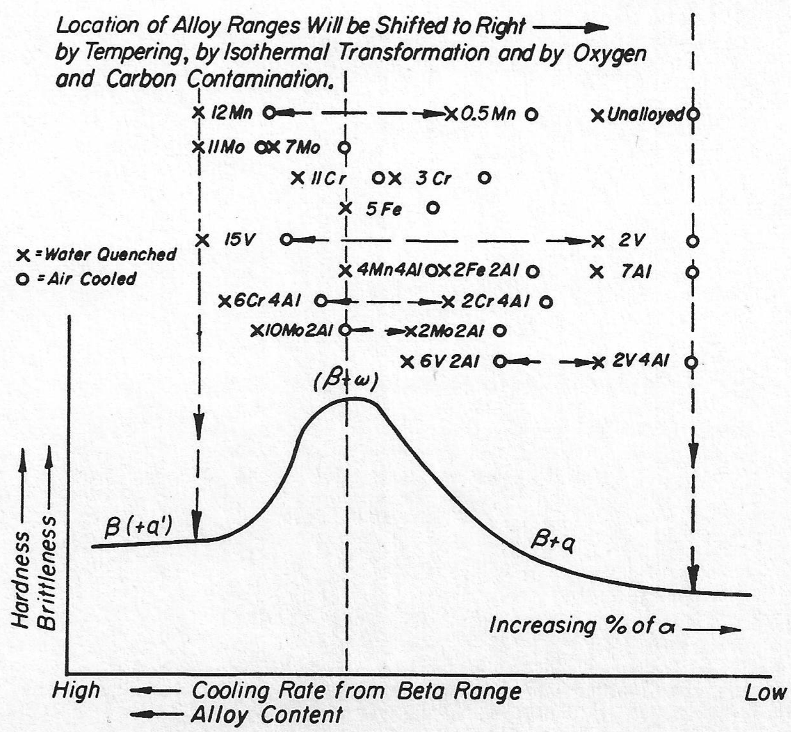
X1300 MOD KROLL ETCH

DISC FORGING

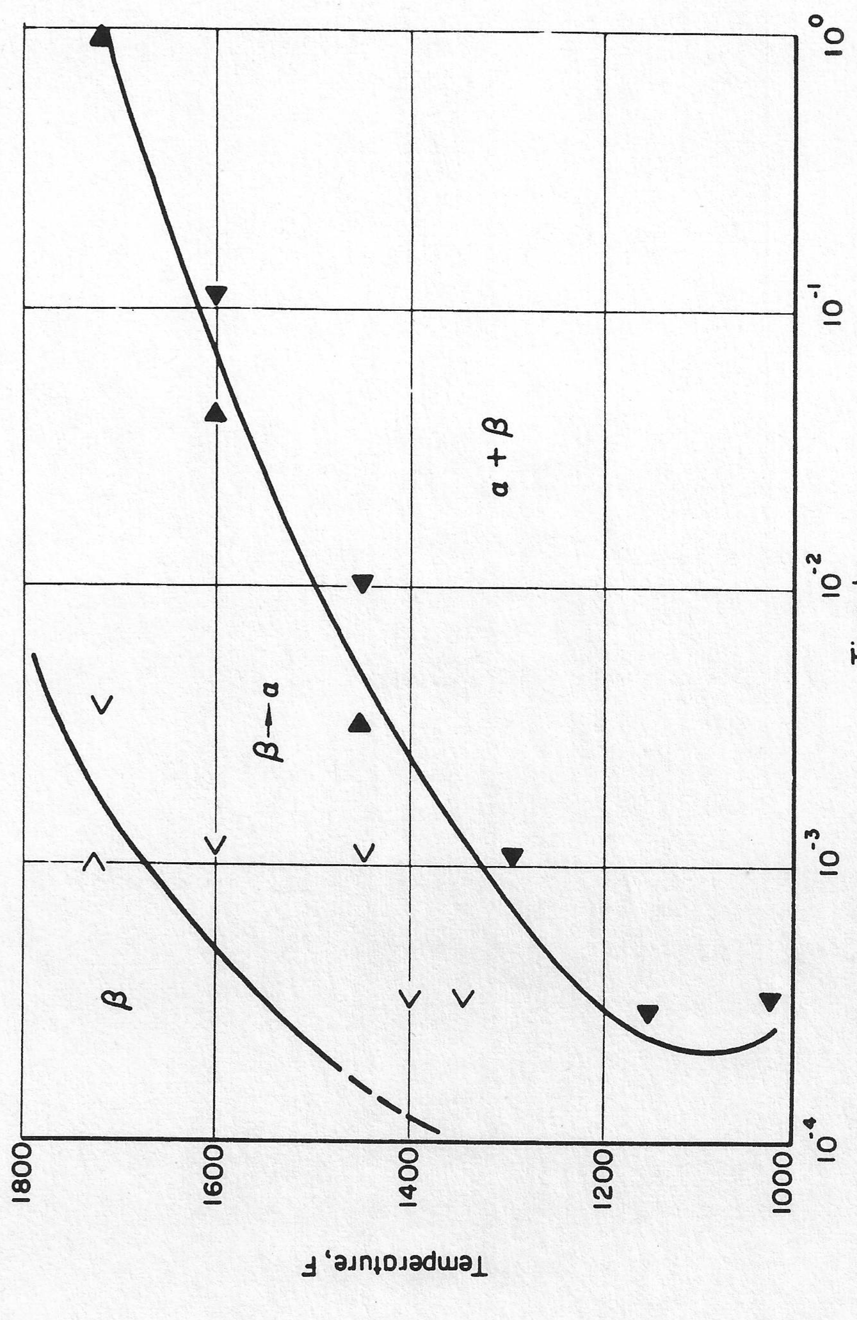
4Al – 4 Mn

1450° F/2 hours, W.Q. and 900° F/8 hours age

Contains primary α islands (appearing mottled due to staining) and an irresolvable precipitate in β.



SCHEMATIC PRESENTATION SHOWING EFFECTS ON QUENCHING FROM BETA FIELD.



TTT DIAGRAM FOR RC-130B (Ti-4Al-4Mn) HELD ONE HOUR AT 1835F PRIOR TO ISOTHERMAL HOLDING
Method: Metallographic
Base: Sponge Ti

At some temperature approaching the upper end of the alpha-beta field (1600 degrees F for 4Al-4Mn; 1450/1500 degrees F for 6Al-4V) the yield/ultimate strength ratio on quenching is at a minimum (approximately 0.50 and 0.70 for 4Al-4Mn and 6Al-4V respectively), corresponding to the temperature of optimum partitioning of the alloying elements with stable beta on cooling. This phenomenon, i.e. the large spread between yield and ultimate strength, is conducive to ease of forming and offers interesting possibilities for straightening operations on blade forgings prior to the final heat treatment.

1.4 Embrittlement due to Forging

Embrittlement (low ductility), can also occur through improper or inadequate forging control quite apart from that due to omega.

Before we were on to the creep stability test, in fact, on our first shipment of 4Al-4Mn compressor discs early in 1954, when we were carrying out laboratory prototype examination with all the enthusiasm of the inexperienced, we encountered a disturbing metallographical condition shown in figure 19. The release notes and control records accompanying the shipment were, as one usually finds in such cases, all in order and very correct. Twenty degrees C tensile test results, the customary control test, all showed excellent ductility values. Discussions with the vendor suggested good forging process control and indicated that the recommended finishing temperature, with adequate reduction, had been followed in forging our discs. Although the forgings had been annealed at 1300 degrees F and air cooled, our knowledge of this alloy, with no datum of experience, had not extended to any great understanding of microstructure or, as previously mentioned, creep stability. In any event, there were the release note results - good, sound, reliable quantitative tensile properties.

This is where we first encountered the significance of strain rate in tensile testing. Equipped with a strain pacer and controlling the rate at 0.005 inch/inch/minute, our tensile elongation values were contrary to the vendor's results, with a wide scatter and many values (4 to 6 percent) below the specified 10 percent minimum. Further work on psuedo forging cycles yielded the microstructure shown in figure 20, from a blade forging specimen heated to the beta field, 1750 degrees F and air cooled. Considering the microstructure of this specimen it will be appreciated that with slightly higher temperatures, or longer times, the residual alpha islands and small grains would disappear, the grain size then approaching that obtained in the disc forging. At temperatures

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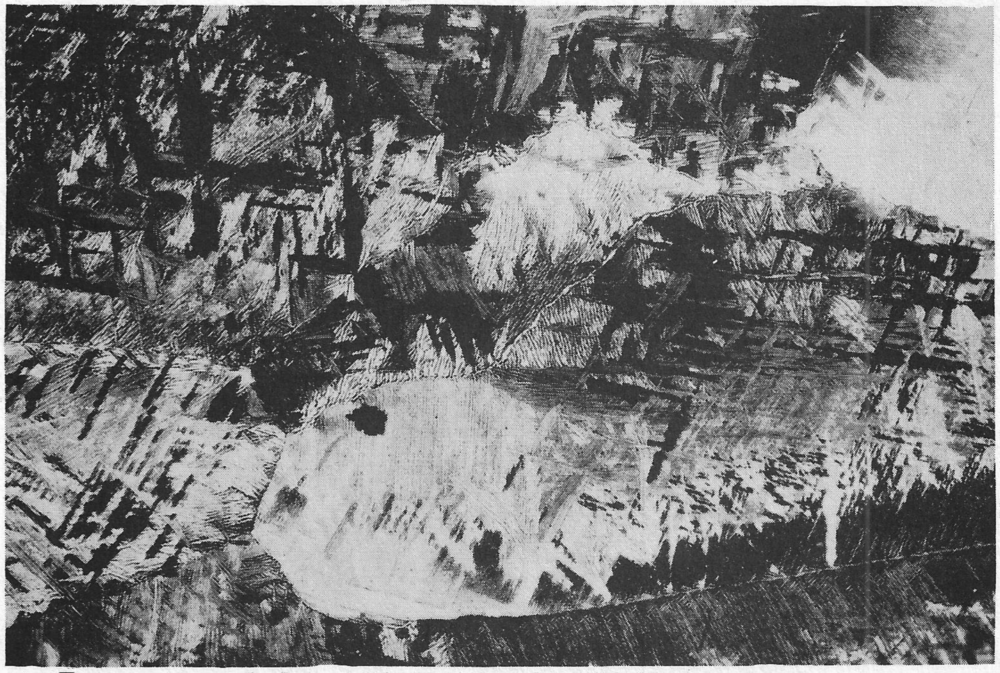


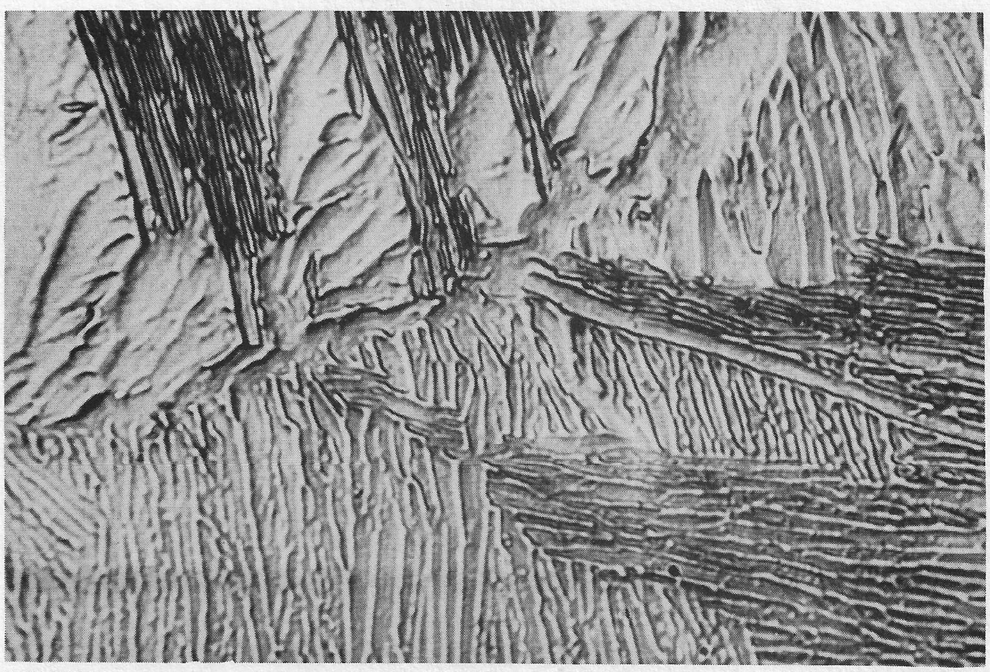
X100 MOD KROLL ETCH

4Al - 4Mn Alloy - Vendor B

COMPRESSOR DISC FORGING

Widmanstätten Structure Alpha Needles In Beta Matrix —

Note Alpha Grain Boundary and Coarse Grain Size —



X1300 MOD KROLL ETCH

Fig. 19

X100 MOD KROLL ETCH

4 Al – 4 Mn Alloy - 1750°F/$\frac{1}{2}$ hour, Air Cool

MARTENSITIC STRUCTURE - Note retained alpha islands (below), mixed grain size (above) and α' needles formed on cooling.

X1300 MOD KROLL ETCH

of this order and with the very slow cooling rate which would be associated with the mass of the disc, one would then expect the alpha dilineated grain boundaries and the basketweave type of Widmanstatten structure shown in figure 19,

In short, the records notwithstanding, we knew that the reduction during the final forging operation had been too low (approaching 10 percent) and the finishing temperature too high (approaching 1750 degrees F) to break-up the structure; this situation had been allowed to exist because the vendor's laboratory technique, also in error, had masked it, and the shop control had failed to ensure conformance to work orders. Subsequently, we were privileged to pass approval on what we consider satisfactory material: i.e. an equiaxed structure of primary alpha and beta, figure 21.

The type of embrittlement discussed here is not to be confused with the omega embrittlement of Paragraph 1.3. The embrittlement caused by heating or finishing at too high a temperature is known as beta embrittlement and will occur when this alloy is heated or finished forged above 1600 degrees F (figure 11 indicates 1650 degrees F for reheating; actually 1750-1800 degrees F). It is not fully understood as yet, but is believed to be due to a combination of the following:-

(a) coarse grain size,

(b) continuous alpha grain boundaries with a Widmanstatten or martensitic structure,

(c) the presence of impurities at the grain boundary.

This phenomenon demonstrates the necessity of finish forging alph-beta alloys, with adequate reduction (about 50 percent plus) at the lowest possible temperature (less than 1650 degrees F) in order to break up the undesirable structure otherwise resulting. The presence of primary alpha islands formed on forging in the alpha-beta field ensures a more ductile structure with less response to aging, i.e. greater stability for subsequent service at elevated temperature.

From theoretical considerations, it might seem possible to break-up such undesirable structures due to forging, by a suitable thermal treatment such as reheating to the alpha-beta or the beta field, furnace cooling or isothermally annealing for long periods of time (10 to 40 hours). In practice the reaction is too sluggish and must be considered impracticable. Only re-working with adequate reduction at a lower temperature

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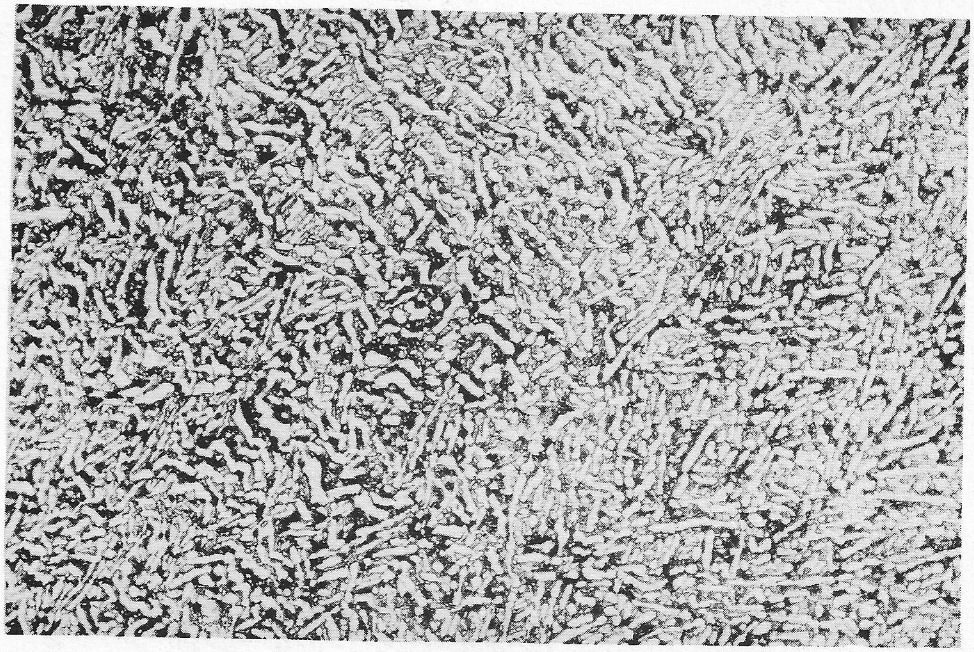
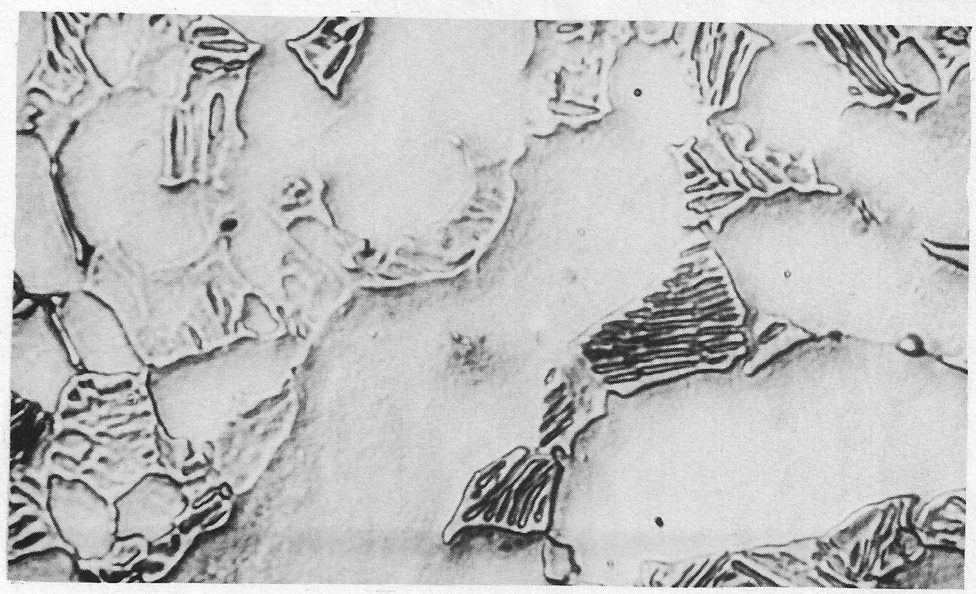


X 100 MOD KROLL ETCH

COMPRESSOR DISC FORGING

4 Al - 4 Mn Alloy-Vendor A

Primary Alpha and Beta Equiaxed Structure – note primary α islands (from forging) and lamellae beta with secondary alpha (formed in annealing)



X 1300 MOD KROLL ETCH

can correct the condition. Here a word of caution must be injected for, carried to the other extreme, excessive reduction at extremely low temperatures can also lead to poor ductility from a fibrous structure which does not respond to annealing.

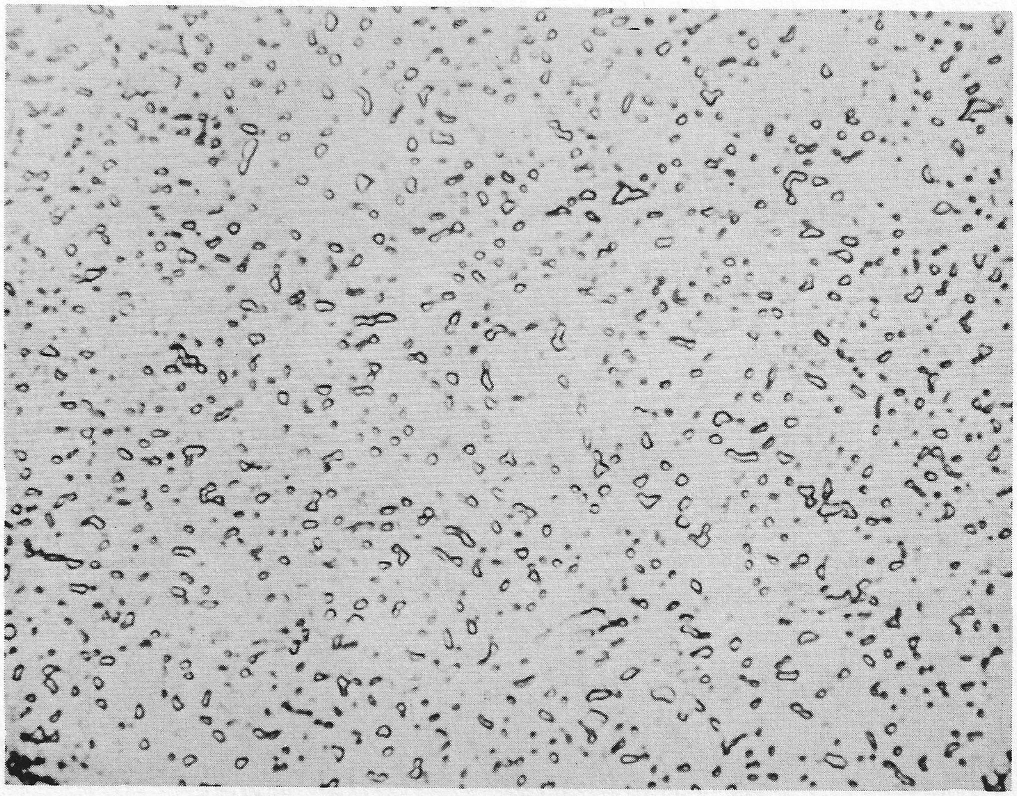
Currently under investigation is a very interesting problem involving a high creep strain value (0.90 to 1.0 percent) obtained on specimens cut from blade forgings and subjected to a stress of 50,000 psi at 400 degrees C for 100 hours. The creep rate represented by this data is almost twice that expected from the alloy. The microstructure is shown in figure 22 and is considered normal. Test values obtained on the annealed forging stock were somewhat lower due to the greater testing accuracy on the larger specimen but still in excess (0.65 percent) of that expected.

The puzzling aspect of this problem lies in the fact that the prime source in confirming these values has also shown satisfactory creep characteristics on a different lot of rolled bar of the same diameter from the same ingot. A 1500 degrees F furnace cool anneal has been employed in this investigation so that recrystallization should not be a factor. No difference in microstructure could be noticed between the two lots although possibly of significance is the fact that they were rolled at different temperatures. The possibility of omega or inadequate annealing treatment as factors is unlikely in view of confirmatory tests carried out. This is not unanticipated in view of the low percentage of beta present which obviates transformation problems. Until all the facets have been explored, which is not the case at this date, no satisfactory explanation can be given.

2.0 FORGING SKIN DEFECTS

The manufacture of jet engine components calls for close quality control in all stages of processing, from melting to the finished part. In certain critically stressed components, absolute reliability is the requirement; the disastrous effect of one defective compressor blade in a compressor containing 1800 blades can be visualized.

In common with other metallic materials, titanium will exhibit defects originating during forging such as cracks, laps, forged-in foreign matter, etc. We have already noted (Paragraph 1.4) that there are sound metallurgical reasons for low finishing temperatures and high reduction in forging titanium. In changing from the



X1300 MOD KROLL ETCH

BLADE FORGING

6Al-4V Alloy

— Beta in Alpha Matrix 1300°F A.C.anneal

Compare the amount of beta with that of 4 Al-4Mn Alloy (Fig.12)

Fe-Cr-Mo Alloy to the 4Al-4Mn material, the relative flow characteristics were clearly demonstrated. Figure 23 is a photograph of the surface of a finished forged blade, 4Al-4Mn Alloy, after etching in 20 percent HNO_3/2 percent HF bath. This type of defect is particularly insidious since the cracks (0.003 inches deep, 0.025 inches in length) are not readily revealed by normal crack inspection methods without prior polishing. The sorting problem was troublesome and it was some relief to find that the introduction of an extra forging operation eliminated the trouble.

The affinity of titanium for oxygen (above 1300 degrees F), nitrogen (above 1500 degrees F) and hydrogen (above 500 degrees F in the absence of oxidizing conditions) precludes long soaking periods or high temperatures. Further, the embrittling effect of carbon dictates a moderately oxidizing furnace atmosphere. The visible oxide scale, formed during forging is not considered a problem since as with other metals, it is readily apparent and easily removed by sand blasting and pickling. However, the subscale or diffused layer requires special attention in view of the above noted affinity of titanium for the interstitial elements.

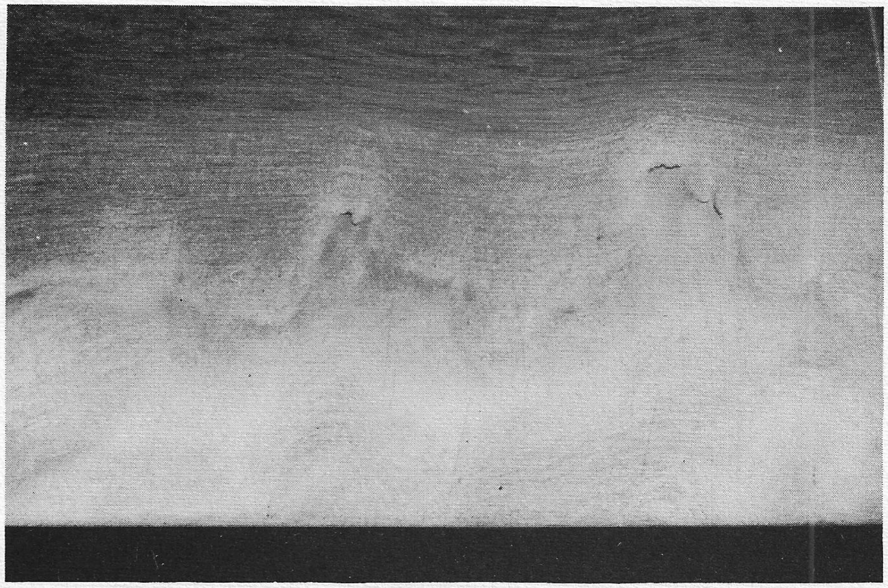
The depth of the scale plus the brittle diffused layer (primarily due to oxygen), has been reported for C.P. Titanium, as follows:-

C.P. Titanium

30 Min. at (degrees F)	Total Affected Layer (inches)
1600	0.0013
1700	0.0022
1800	0.0038

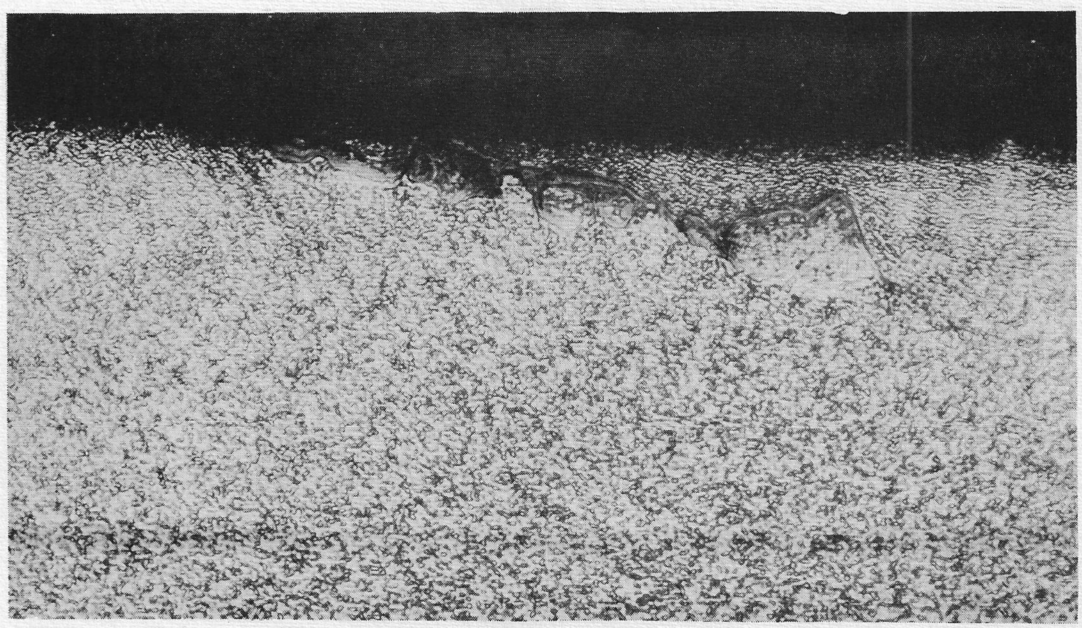
Oxygen penetration follows the normal parabolic diffusion law so that at 1700 degrees F, for example in alpha-beta alloys, a depth of about 0.015 inches is produced in one hour, but four hours produces only 0.030 inches. At longer times, the loss of surface material, as scale, tends to keep pace with the growth of the subsurface layer so that the latter does not increase with time. Four hours at 1500 degrees F will produce 0.002 inches while at 1300 degrees F, no diffusion is detectable and the thin oxide scale may be removed by cleaning.

On precision forgings, usually, sand or grit blasting plus polishing is sufficient to clean up, (the order of magnitude of the diffused layer will be 0.001 inches or less). Since a thin silicon layer may be retained from the sand blasting operation, forming titanium silicide (Ti_5S_3) on annealing, it is usual to follow



X3 1/2

Blade Surface After Etching In HNO_3/HF



X200 MOD KROLL ETCH

Microsection Through Forging Crack

4Al - 4Mn ALLOY

Fig. 23

blasting with a nitric-hydrofluoric dip regardless of the amount of polishing to follow. On heavier forgings, the diffusion zones are greater due to the longer times and higher temperatures employed and such forgings should therefore be machined on all faces.

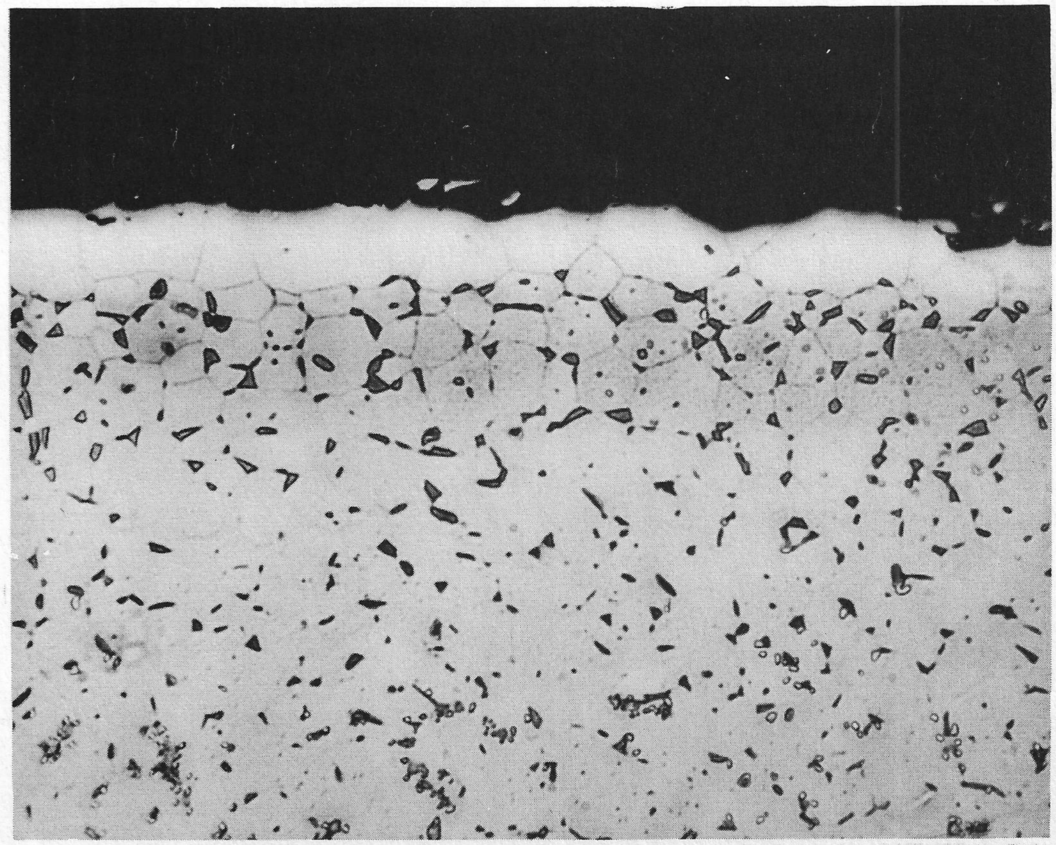
Unlike oxygen and nitrogen, hydrogen diffuses readily throughout the mass at normal hot working temperatures and can only be removed by vacuum treatment. With careful attention to processing details hydrogen absorption during hot working can be minimized at innocuous levels (current shipments are less than 70 to 90 ppm) but may creep in from other sources.

For instance, figure 24 shows the surface of a blade forging in an all-alpha Alloy, 5Al-2½Sn. The photomicrograph is notable in several respects i.e. it shows a single phase (alpha) skin, a dark etching phase just below this surface decreasing and associated with small islands of beta phase towards the centre. The single phase alpha at the surface is attributed to the effect of oxygen diffusion; the clear islands of beta towards the core are normally encountered in this alloy with an iron content of 0.15 to 0.20 percent. The origin of the dark etching phase was not clear at first.

Chemical analysis revealed excessively high iron (0.69 percent versus 0.50 percent specified) which with hydrogen is a potent beta stabilizer. Then, it was found that this dark etching phase near the surface would disappear on vacuum degassing, indicating hydrogen as a factor. Actual analysis showed a hydrogen content of 96 ppm at the centre and 181 ppm at the surface (0.005 inches deep) suggesting the possibility of even higher hydrogen contents at the outer 0.001 inches/0.002 inches. Normally, rapid diffusion of hydrogen takes place at the annealing temperature (1500 degrees F), but, since this did not occur in this instance, it was concluded that hydrogen contamination occurred during the final acid de-scaling bath. Such de-scaling or light pickling baths e.g. the 25 percent nitric/2 percent hydrofluoric bath at 150 degrees F must be so balanced and maintained that no hydrogen pick-up will occur. Fused caustic baths (85 percent NaOH, balance oxidizing agents) such as Virgo or Kolene at 900 degrees F are also satisfactory for descaling operations but the sodium hydride type of de-scaling bath is a potent source of hydrogen pick-up and should not be used.

The alpha skin noted above and attributed to oxygen and possibly nitrogen is of course brittle due to the influence of these elements. It can occur with any alloy and if present in the finished part will lead to premature failure.

Much attention has been directed towards the use of nickel plating as a protective coating for precision forgings, to minimize



X 640 MOD KROLL ETCH

5 Al - 2 1/2 Sn BLADE FORGING

Anneal 1500°F/ 1/2 hour, A.C. - note alpha skin, gradation of dark etching beta phase (Fe plus H_2)and clear beta islands (bottom) skin effects due to sub-scale diffusion. The technique relies on the simple cylindrical shape of the stock, compressive stresses in the plate, and a certain degree of bonding (diffusion of nickel) from the forging operation in order to obtain the necessary adherence and protective value. While largely successful in achieving the objective of eliminating oxygen diffusion. nickel itself readily alloys and will diffuse inwards (figure 25). Nickel penetration therefore, must be controlled, like oxygen, to a depth of less than 0.001 inches to ensure removal by grit blasting. Nickel plating may offer certain advantages in that is reduces the risk of hydrogen absorption, provides protection against mechanical damage and may possibly act as a lubricant during forging.

3.0 MELTING DEFECTS

Quality problems originating at the melting stage are usually related to either contamination or segregation (chemical inhomogeneity). Contamination is illustrated in figure 26 showing tungsten carbide inclusion.

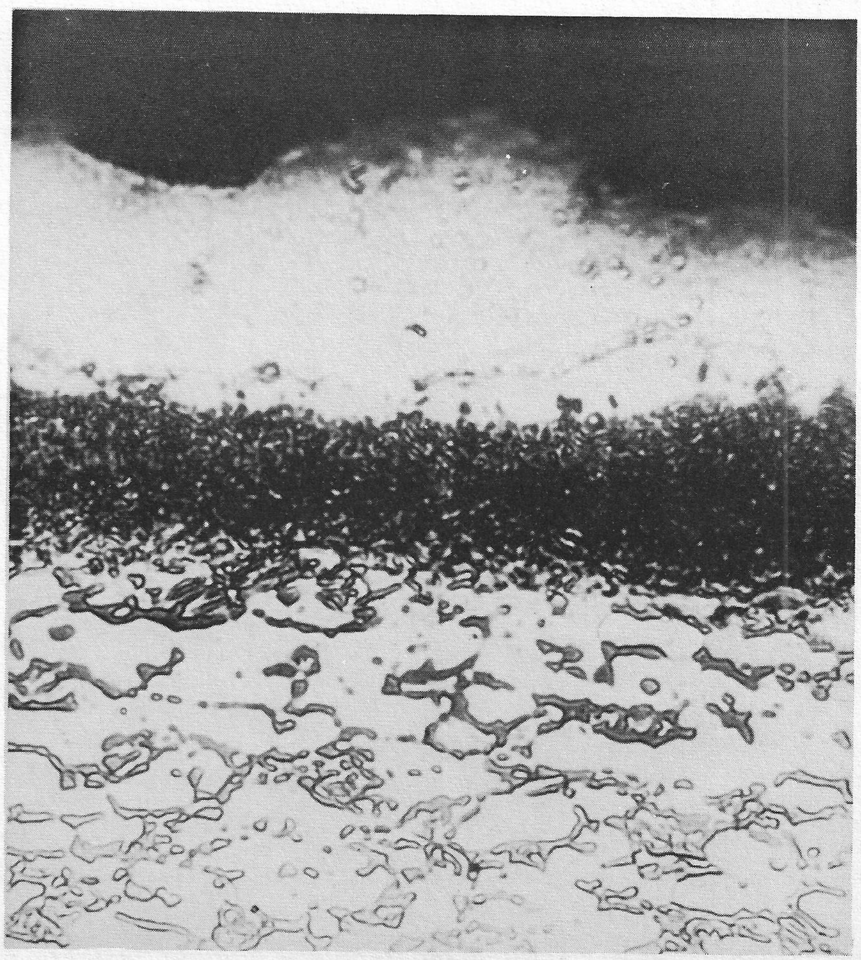
Due to its high density, tungsten or tungsten carbide is readily detectable by radiographical methods. Our previous (1951) experience had shown the tungsten tip electrode, used in arc melting, as the contaminating source but since melting methods now employed the consumable electrode technique, the tungsten carbide could only be attributed to:

(a) trace particles from the tungsten electrode welding process used in joining the electrodes together for the second melting stage, and/or

(b) the re-cycling of primary scrap material where turning of the ignot or billet had been carried out with tungsten carbide tools.

The latter proved the case; careful visual and magnetic separation of any scrap used has prevented recurrence of the trouble.

During this period, the prime source did not hesitate to inform us that they had not received complaints from their other customers (they did in due course) and that, since we tolerated inclusions in steel, we were discriminating against titanium. Our concern on the latter point lay in the widely different properties between the titanium matrix and the extremely hard tungsten carbide particles, aside from any consideration of the fact that tungsten is not indigeneous to titanium melting practice. Finally, the associated cracks, shown in figure 26 brought the point home



X1300 MOD KROLL ETCH

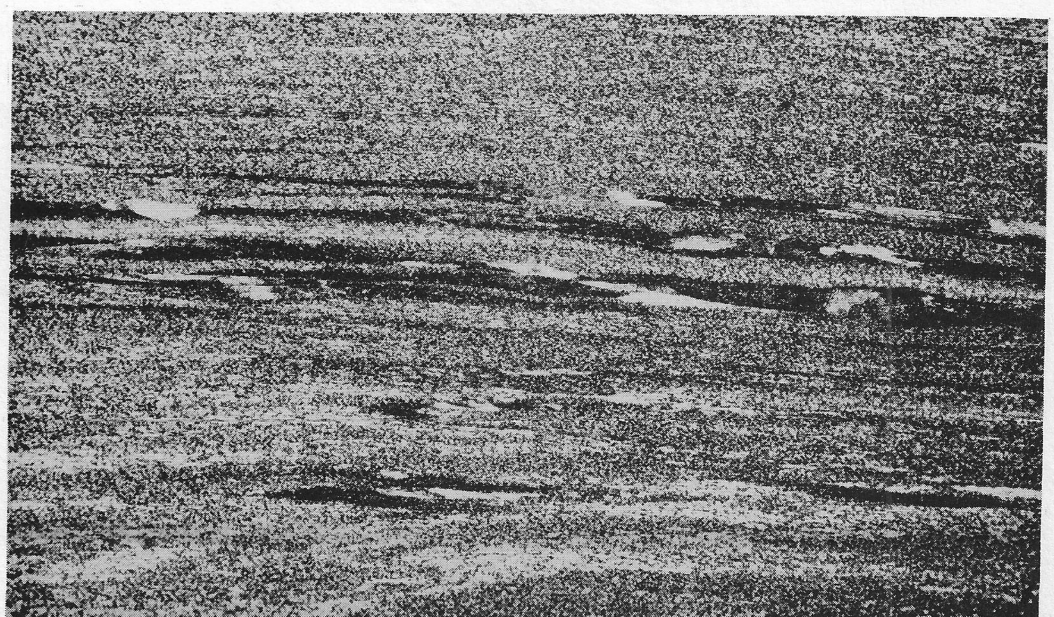
NICKEL PLATED 4Al-4Mn BLADE FORGING

As forged and annealed. Note surface alloying beneath nickel plate.

X500 AS POLISHED

Tungsten Carbide Inclusions In 4Al-4Mn Alloy —
Note associate cracks

Fig. 26



X 100 MOD KROLL ETCH

Banding Segregation In 4Al-4Mn Alloy

Fig. 27

and dispelled any further discussion on their deleterious nature. The problem of the presence of beta phase in an all-alpha alloy, to which previous reference has been made, see figure 24, falls in the same category. On analysis, iron, a beta former was found to be sufficiently high to form a duplex structure.

Freezing segregation, resulting in chemical composition gradients is always a potential problem. Consequently, we were reasonably sure that we had a segregation problem when we encountered the two conditions, banding and peripheral, shown in figures 27 and 28. The forging vendor was of course pleased to find the same condition in his forging stock, figure 29, thus pointing the way to the prime source. This condition could only be seen at low magnification (less than 10X); at high magnification, no difference in microstructure could be noted between the centre and edge. The important questions, however, were concerned with the cause and influence on properties.

Utilizing the spectrographical micro analysis technique of the Department of Mines and Technical Surveys, Ottawa, Ontario, manganese was readily determined as the offending element with variations as high as $\pm$ 25 percent of the average reading. Chemical analysis confirmed the manganese variation with values of 3.6 percent being obtained on outside turnings (0.125 inches deep) versus 4.1 percent at the centre. The difference in manganese content between the surface and centre caused a fine grain size at the outside and coarse grain size at the centre for the same rolling temperature; hence the difference in macrostructure and not microstructure.

Manganese has a pronounced tendency towards freezing segregation (more so than vanadium) which encourages the chemical gradient noted. Due to its volatilization characteristics, an argon sweep is frequently employed during melting.

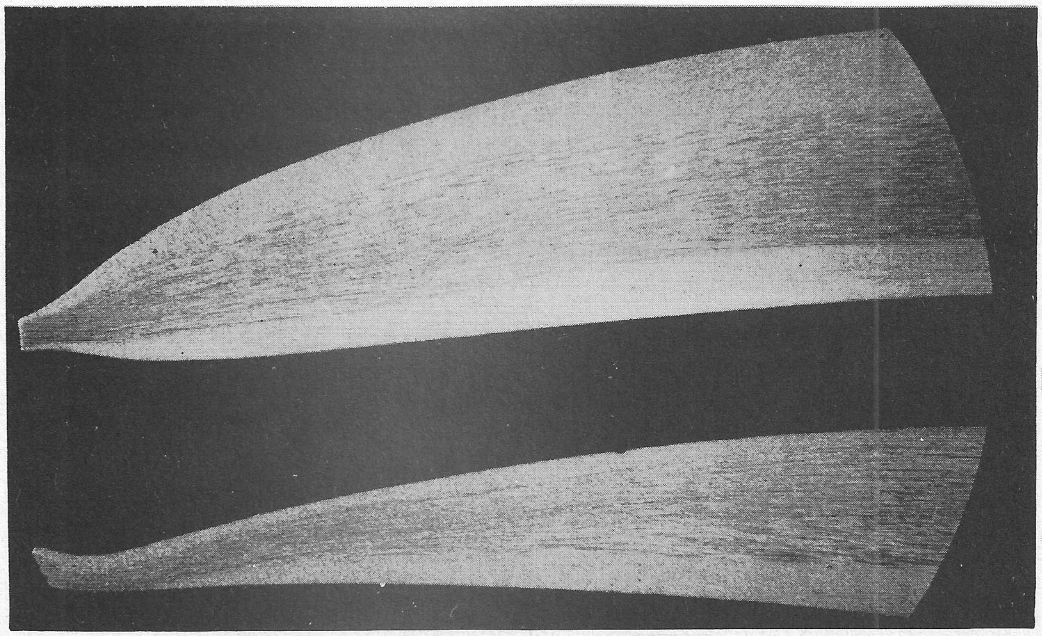
Specification limits for both aluminum and manganese are three to five percent (nominally four percent for both elements). The individual effects can be shown by the following data:-

TABLE V

The Effect of Aluminum and Manganese on Properties

Al (percent)	Mn (percent)	UTS (psi)	0.2% YS (psi)	El (percent)
4	3	148,000	140,000	14
4	4	148,000	140,000	14
4	5	148,000	140,000	14

ORENDA

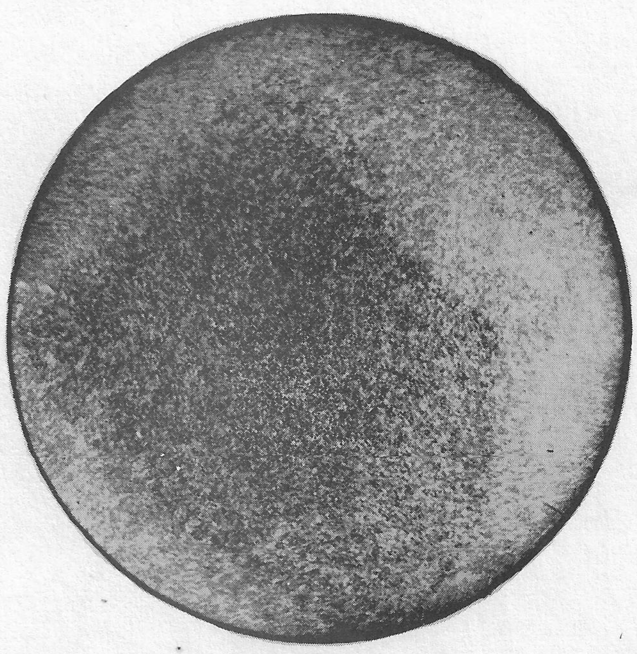


X 6 KROLL ETCH

4Al-4Mn ALLOY BLADE FORGING

Note peripheral banding

Fig. 28



X 6 KROLL ETCH

CROSS SECTION OF FORGING STOCK

11/16 inch diameter, hot rolled round

Fig. 29

TABLE V con't

Al (percent)	Mn (percent)	UTS (psi)	0.2% YS (psi)	EL (percent)
3	4	140,000	130,000	15
4	4	150,000	140,000	13
5	4	160,000	152,000	10

Fortunately, there is very little tendency for aluminum variations or composition gradients during melting.

4.0 WELDING AND BRAZING

In many respects, the welding procedures for C.P. titanium and titanium alloys are not unlike those for other metals, particularly stainless steels. Consideration of the physical metallurgy of titanium as discussed in the preceding chapters, will indicate however, certain precautions necessary in welding this group of metallic materials if brittle welds are to be avoided. Indeed, metallurgical considerations dictate the types of weldable alloys and the welding methods as well for although titanium and its alloys are "weldable" in the sense that cracks may not occur in the as-deposited metal such welds may possess low ductility and be susceptible to cracking by very low thermal or mechanical stresses.

A welded joint has four distinct areas of interest viz. (a) the cast weld metal; (b) fusion line; (c) heat affected zone wherein the base metal has been subjected to various temperatures from the melting point down with very fast cooling rates, and finally (d) the unaffected base material. Hence, we are concerned with those problems related to casting and heat treatment of titanium.

4.1 Transformation

Alpha-beta titanium alloys, as we have noted, are prone to beta transformation problems of embrittlement or instability in the as-cooled or as-quenched state. Such alloys are hardenable by either of two mechanisms i.e. by martensitic shear reaction at fast cooling rates or nucleation and growth at slow cooling rates. A wide variety of microstructures, all brittle, are produced in cooling from high temperatures. Hence, alpha-beta alloys such as 4Al-4Mn, in general, are not recommended for welding. Theoretically possible post-welding heat treatments involving reheating to beta or alpha plus beta field with or without furnace cooling or isothermal annealing are not practical for sheet metal components due to the danger of contamination and scaling during the very long periods at the high temperatures involved.

In comparison with inert gas metallic arc welding, problems associated with transformation are reduced but not entirely eliminated with the lower heat input and faster thermal cycle of resistance seam or spot welding. With this welding method, some degree of success has been reported in welding the 8Mn alpha-beta alloy and with the 6Al-4V alloy which contains only a small quantity of beta in its normal structure.

In flash welding, pressure upsetting expels the molten metal and hence any brittleness associated with freezing segregation or other undesirable structure in the cast weld metal is absent. This process is widely used for making rings, particularly stator rings for jet engines. A microstructural study of such a weld in 6Al-4V alloy is shown in figures 30 and 31. The heat affected zone is distinguished by undissolved islands of alpha and the fusion line by a Widmanstatten type of structure typical of slow cooling. Mechanical properties across the joint match parent metal properties with the exception of ductility which is partially influenced by the transverse grain flow at the interface due to the upsetting and expulsion of the metal in this area.

The foregoing will explain why the all-alpha materials (5Al-2½Sn and C.P. Titanium) are preferred for welding applications regardless of method. Such alloys are not subject to transformation problems i.e. heat treatment response. Grain growth does not seem to have an effect on properties and very satisfactory properties are realized.

TABLE VI

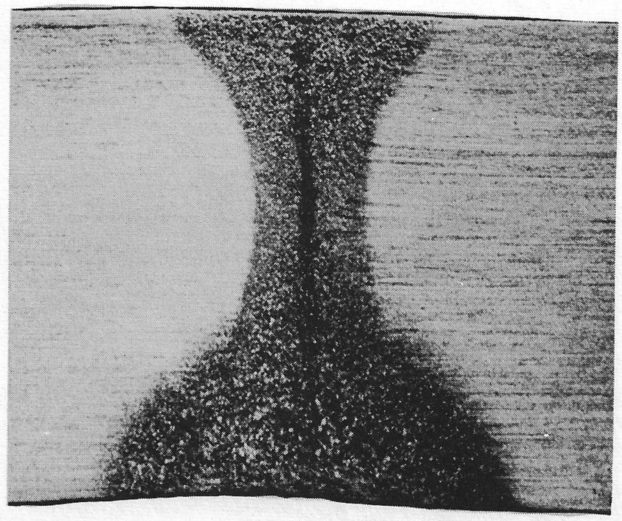
Average Mechanical Properties of Welds in Titanium Alloy

Alloy	Welding Method	UTS (psi)	0.2% YS (psi)	El (percent)	R.A. (percent)
6Al-4V	F.B.W.	153,000	147,000	12	40
5Al-2½Sn	SIGMA	131,000	122,000	16	32

The welding of titanium to other metals is not possible due to the formation of brittle compounds between titanium and high percentages of the other elements, as indicated in the phase diagrams.

4.2 Contamination

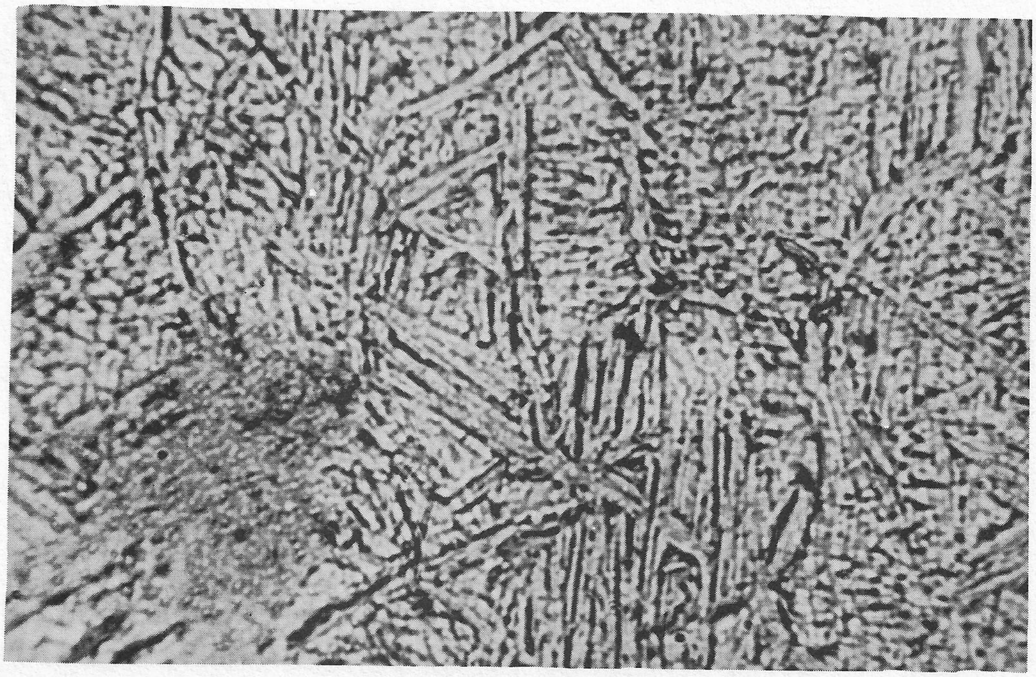
Welding methods, procedures and techniques are limited by



X6 MOD KROLL ETCH

6AI – 4V ALLOY - FLASH BUTT WELDED

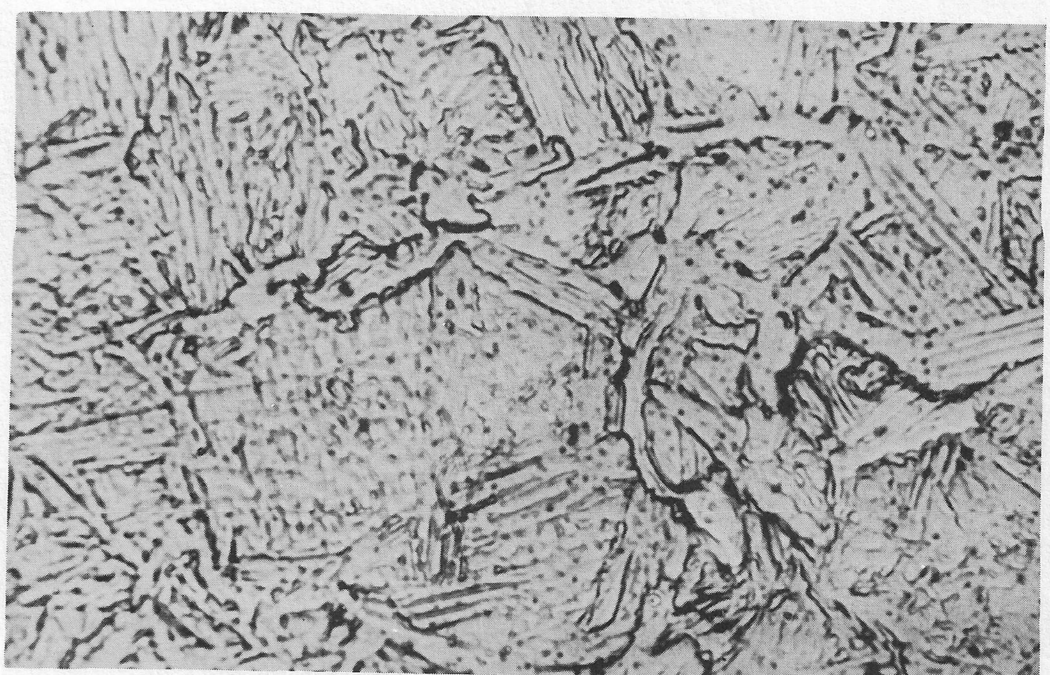
7/16 inch x 1 1/2 inch x 32 1/2 inch O.D. Ring



X1300 MOD KROLL ETCH

6Al-4V ALLOY-FLASH BUTT WELDED

Fusion line (above) and heat affected zone (below) – Parent metal is similar to that shown in Fig. 22



X1300 MOD KROLL ETCH

requirements for the prevention of weld contamination by the interstitials, oxygen, hydrogen and nitrogen in the atmosphere (or carbon from unremoved oil or dirt on the part). Inert gas shielded metallic arc with consumable or with added filler rod, spot- seam- flash- or pressure welding are the only methods which offer any degree of protection against these impurities at the molten weld metal and heat affected zone.

Little or no difficulty is encountered in welding titanium by resistance methods, but ingenuity in setting-up and careful attention to details are necessary if contamination is to be avoided when welding manually or automatically with the inert gas metallic arc method. Welding procedures and techniques are established with the following objectives in mind:-

(a) complete shielding during "open air" welding by using oversize nozzles, baffles, a trailing shield (figures 32 and 33), root shielding backing fixtures and high gas flows (short of the point where the gas actually "rolls in" oxygen due to turbulence). Alternatively, "chamber" welding (figures 34 and 35) may be used especially for materials in excess of 0.10 inches where the high heat input necessary and slow dissipation makes "open air" shielding difficult or where fillet welds call for difficult set-ups.

(b) high purity shielding gas; argon (99.998 percent purity) is preferred for "open air" welding due to its higher density and its stable effect on the arc. However, helium allows higher welding speeds and better penetration.

(c) minimum heat input as influenced by joint design, welding speed and current. The use of steel back-up bars retards the heat dissipation at the weld area and permits lower current settings with less turbulence in the weld pool and reduced contamination from the air. The shape and size of the groove in the back-up bar may be altered to provide various heat transfer characteristics (3T wide by 3T deep is usual). Hold down or chill bars are chamfered and positioned as close to the weld as possible.

(d) precleaning thoroughly is an absolute prerequisite. Depending upon the prior condition, this may be done by either acid pickling or vapour blasting (light scale) or (with heavy scale) mechanical cleaning (sand, grit) or de-scaling in a Virgo bath. Hand polishing of both joint edges and filler rod just prior to welding is desirable but not mandatory.

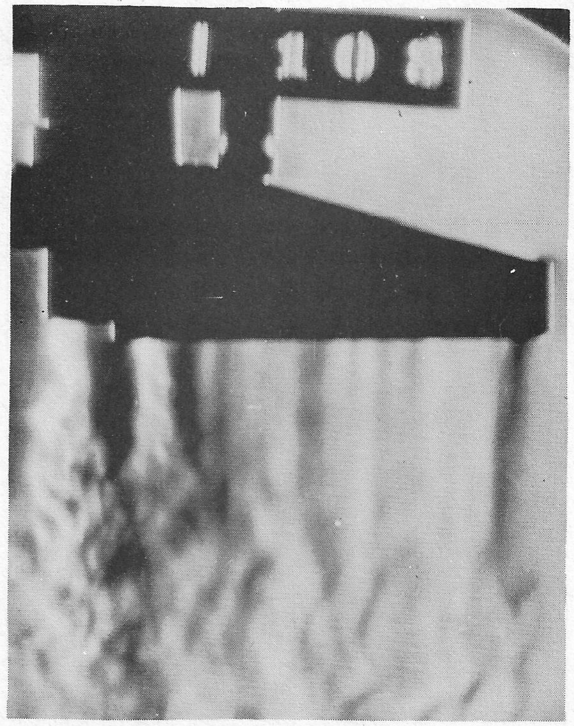
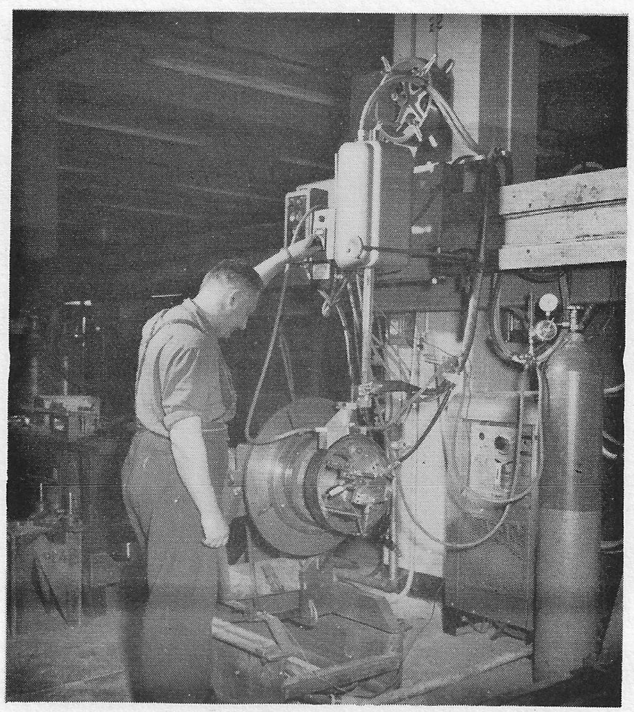


Fig. 32



"OPEN AIR" WELDING WITH TRAILING SHIELD ADAPTOR

Fig. 33

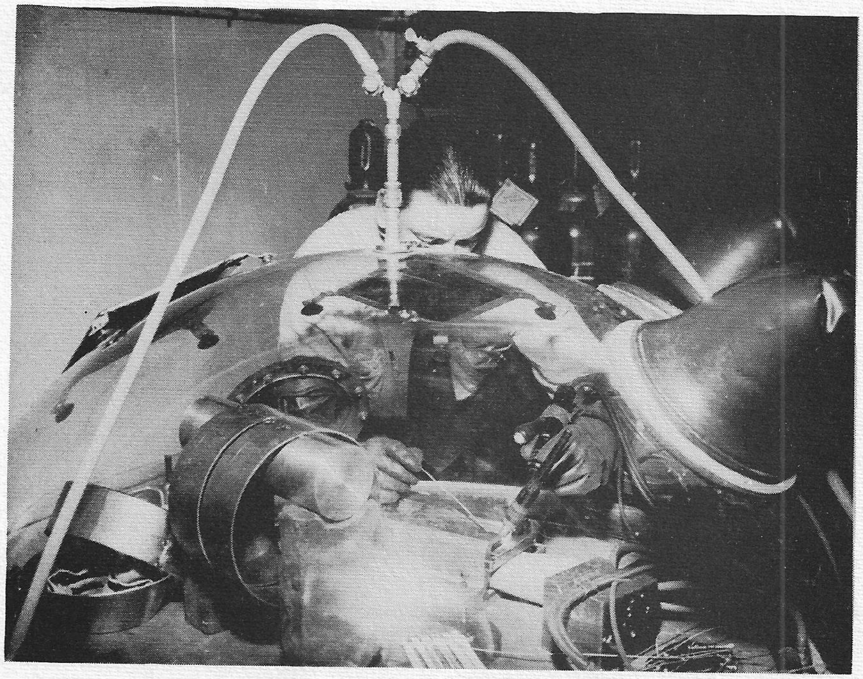
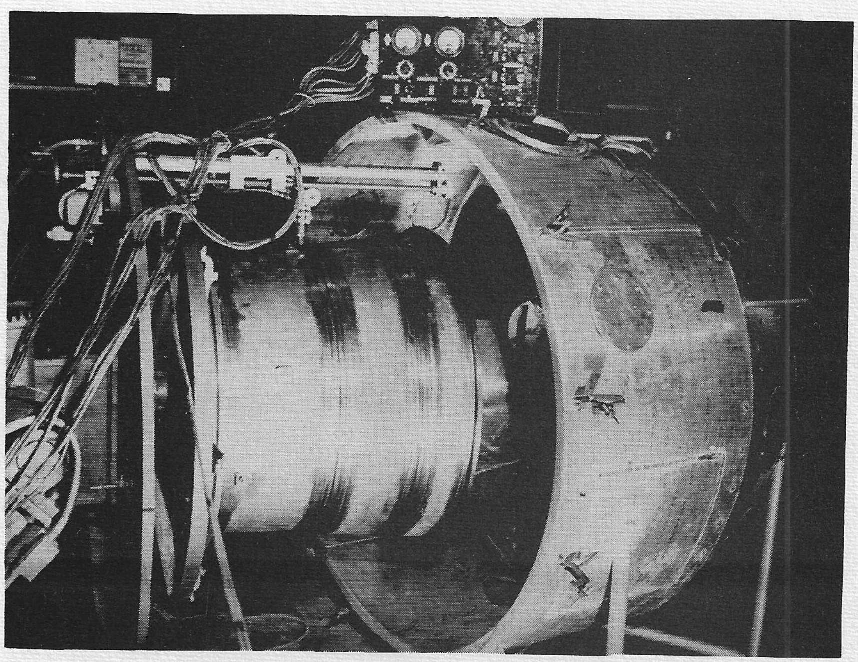


Fig. 34



CHAMBER WELDING WITH ARGON ATMOSPHERE ——

Fig. 35

(e) operator technique when using added filler rod is directed towards short arc lengths and avoiding the withdrawal of the rod from the shielding gas which would contaminate the rod and subsequently the weld. Since added filler rod creates turbulence in the shield and hence contamination of the weld, consumable electrode welding or fusion e.g. butt welding without added filler rod (for sheet thicknesses up to 0.062 inches) is to be preferred.

With good welding practice, the hydrogen content of the weld metal may even show a reduction of approximately 25 per-cent from that predicted by the analysis of the rod or elect-rode and base material. Weld surface is lustrous under these conditions and shows no discolouration. However, the presence of discolouration (straw, blue, etc.) does not necessarily always indicate contamination and hence brittle welds nor do the different colours offer an infallible guide to relative temperatures for the colour is time dependent and may be only a surface effect. It may also be possible to have improper shielding due to turbulence at the arc zone, yet adequate shielding at lower temperatures away from the arc; which conditions could result in a shiny, lustrous appearance and brittle welds. A grey scale condition is the end-point of contamination having zero bend ductility.

4.3 Brazing

Although it has not been developed to a satisfactory level of reliability as yet, the potentiality of brazing using silver or silver manganese brazing alloys, appears very good. Induction brazing in a dry inert gas atmosphere, vacuum furnace brazing and inert gas shielded tungsten arc brazing are methods offering possibilities.

Unlike nearly all other elements, the titanium silver compound is inherently ductile; in addition titanium has low solubility in molten silver which minimizes erosion and under cutting effects during brazing. As with welding, joint cleanliness is of paramount importance.

The alpha type alloy offers additional scope in developing a suitable brazing alloy, since transformation effects are not present in these materials. Experimental work elsewhere, is directed towards the nickel base, titanium, copper, cobalt, series which have flow points in the range of 1750 to 1950 degrees F and may be suitable for furnace brazing once developed.

5.0 MECHANICAL PROPERTIES

The mechanical and physical properties of titanium and titanium alloys have been extensively reported elsewhere and it is not the writer's intention to repeat them here.

Titanium is noted for its low density, low coefficient of expansion, low thermal conductivity, intermediate modulus of elasticity (16×10^6) and its extremely bad galling or fret welding characteristics. This latter property is of great significance in engine application and we will return to it later. It might be supposed that the rapid absorption of hydrogen in titanium, mentioned earlier as increasing rapidly above 500 degrees F (260 degrees C) might be a limiting factor in the application of titanium alloys. Fortunately, as noted previously this temperature is significant only under non-oxidizing conditions. The following data reported by Rem-Cru and subsequently verified by Orenda Engines Ltd., will illustrate the absence of hydrogen pick-up under atmospheric conditions of time and temperatures which are of interest to the design engineer.

TABLE VII

Alloy	Hydrogen as received (ppm)	Hydrogen after 500 hours at 800 degrees F (ppm)	Hydrogen after 500 hours at 1000 degrees F (ppm)
6Al-4V	59	54	62
6Al-4V	76	83	108
5Al-2½Sn	74	76	76
8Mn	179	178	109
8Mn	150	185	165
6Al-4Mo	92	99	97

It is apparent that no significant hydrogen pick up has occurred and this is attributed to the barrier-effect of the thin oxide film formed on heating (to the temperature noted).

The great attraction to the designer is the outstanding strength/weight ratio of titanium alloys in which it is superior to other metals, particularly in the intermediate temperature range of 100 degrees C to 400 degrees C or higher. It is conventional procedure to illustrate this superiority, which is done in figure 36, for a number of alloys of direct interest to us. Since yield strength and creep characteristics at temperature are related to disc and ring applications, the following table, (not corrected for density) will show the relative strengths of the same alloys.

ULTIMATE STRENGTH / DENSITY VS TEMPERATURE

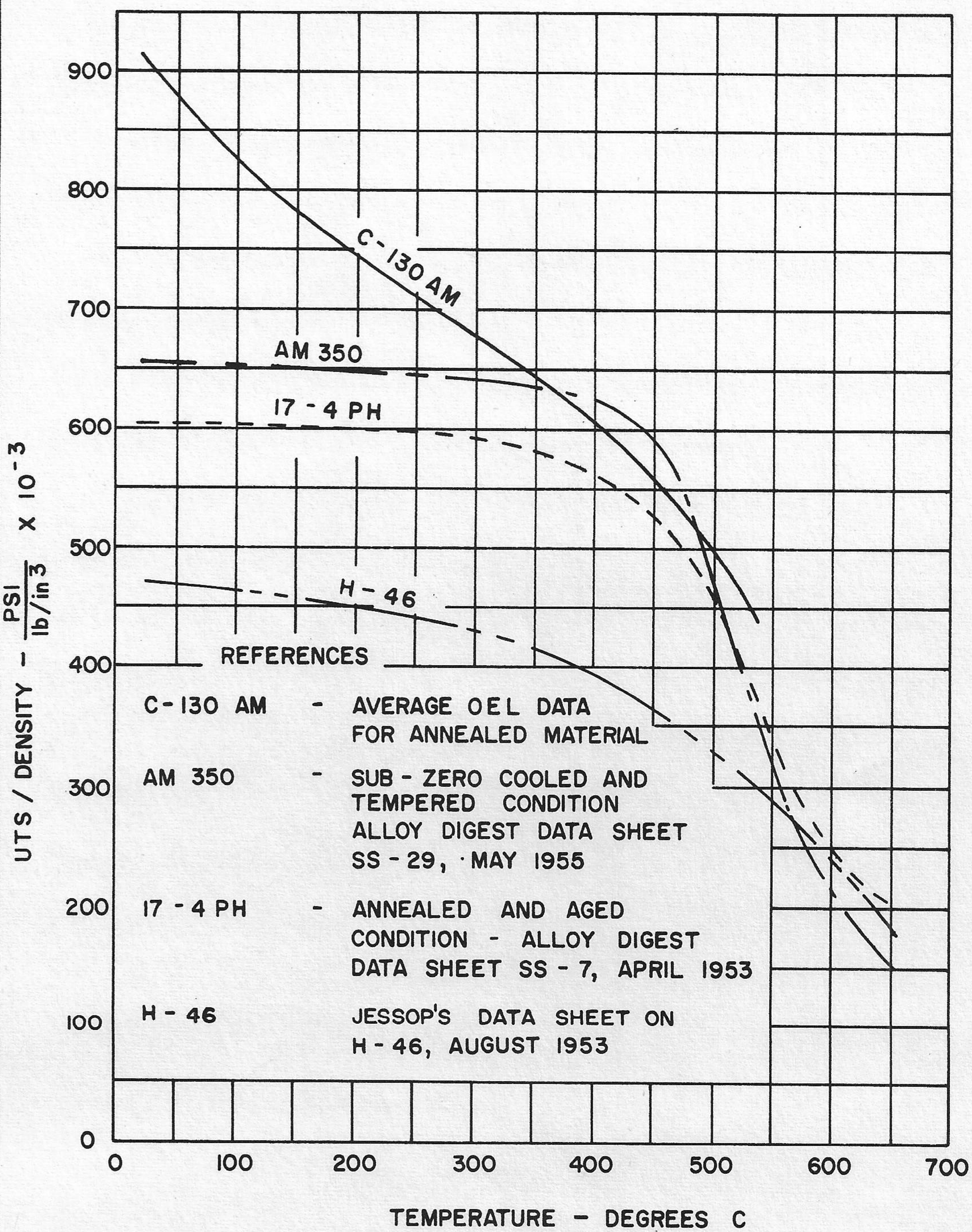


TABLE VIII

Relative Strengths of Various Alloys

Alloy	0.2% YS (psi)	1000 hour Rupture (psi)	0.2% YS (psi)	1000 hour Rupture (psi)
	600 degrees F		800 degrees F	
403	23,000	29,600	-	-
H-46	35,000	-	31,400	-
AM350	45,000	-	40,500	55,000
4Al-4Mn	100,000	95,000	88,000	46,500
6Al-4V	110,000	100,000	100,000	73,000
5Al-2½Sn	67,000	65,000	63,000	58,000
	700 degrees F		900 degrees F	
403	-	-	17,500	14,300
H-46	32,500	-	28,600	(24,800)
AM350	-	-	35,000	34,600
4Al-4Mn	95,000	70,000	68,000	-
6Al-4V	105,000	90,000	88,000	50,000
5Al-2½Sn	65,000	60,000	62,000	40,000

The standard quality control test for bars and forgings is the simple tensile test at room temperature (and the transverse guided bend test for weldable sheet materials). We have already noted the significance of laboratory technique, particularly the control of strain rate, in reporting tensile values. Although the room temperature test is not an infallible guide to elevated temperature stability or notch sensitivity, it owes its wide use to precedent with other metals, simplicity and ease of performance. Such tests should, however, be backed up with periodic notched and creep stability results.

Figures 37 to 44 inclusive are a statistical analysis of a large number of tensile tests on specimens cut from three different types of 4Al-4Mn components, rings, discs and blade forgings from three sources. As such they do <u>not</u> represent idealized metallurgical grain flow conditions but do give the user a much closer picture of what to expect in the actual part. The isolated tensile strength values falling at, or just below specification, particularly on heavy forgings such as rings are not considered significant particularly in view of the grain flow and the low oxygen melt involved; 0.2 percent yield strength and ductility, the properties of greatest concern in such applications, are well up. Since a controlled amount of oxygen is used in this alloy to maintain strength levels, the significance of the scatter shown is

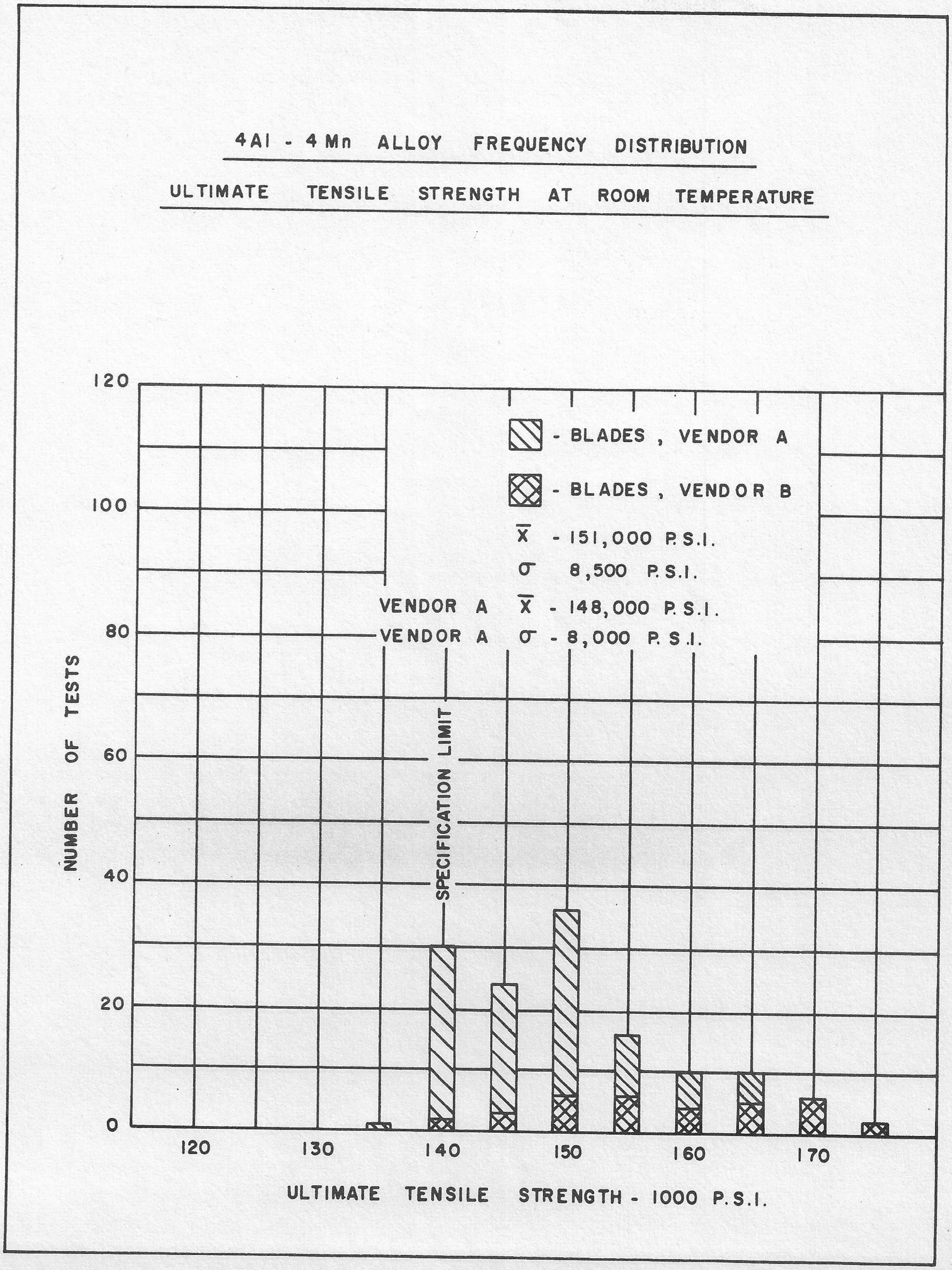
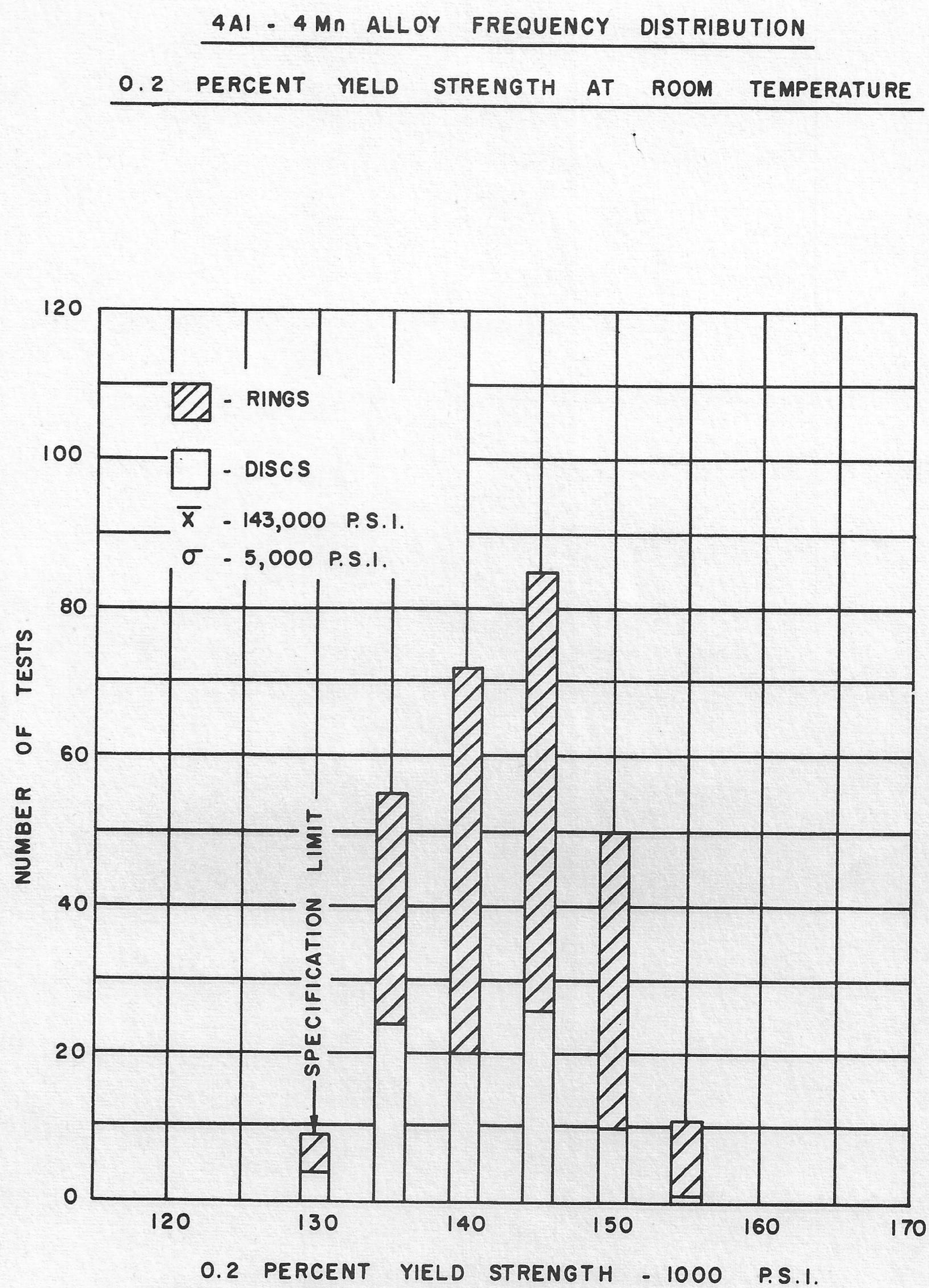
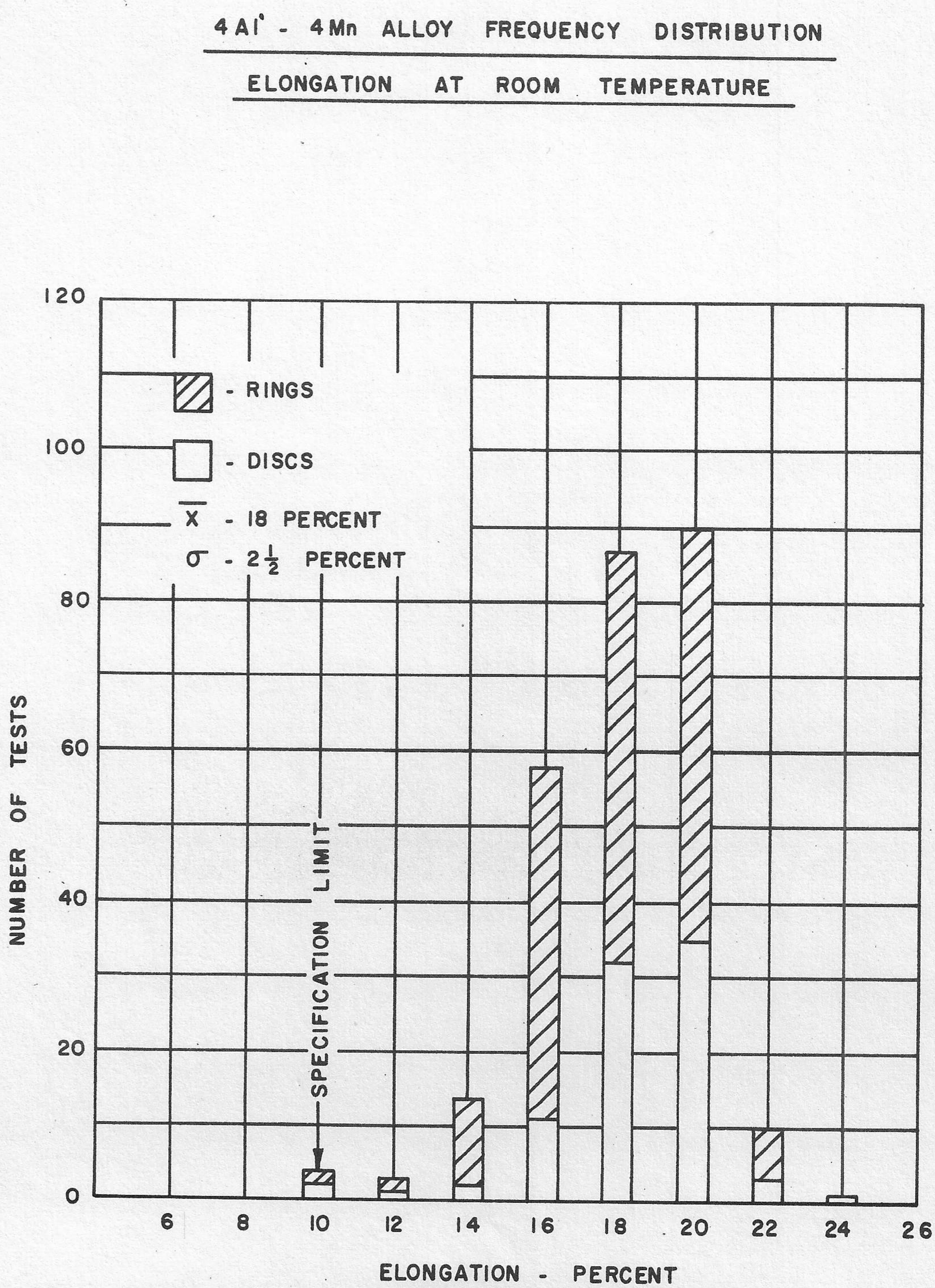
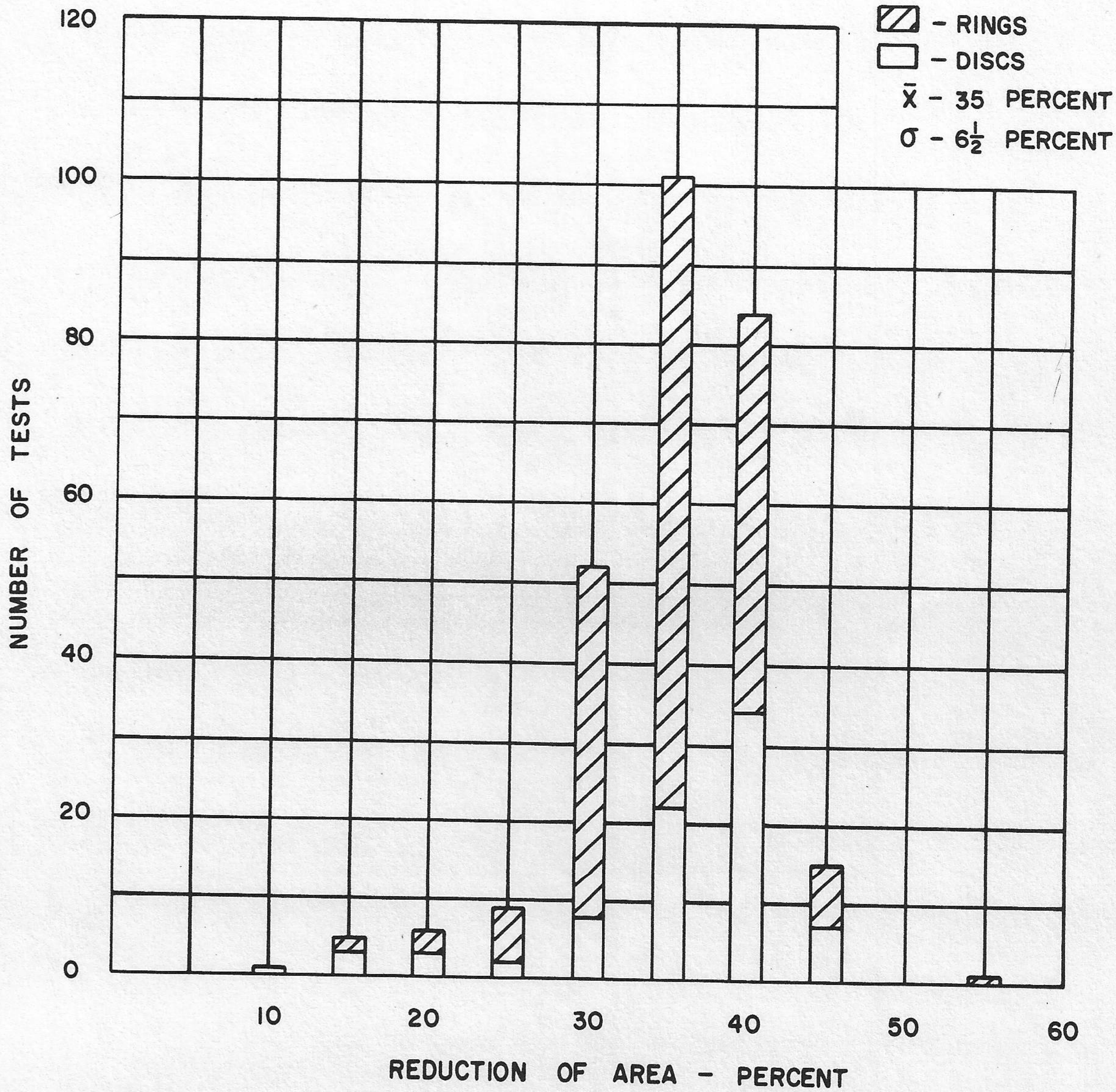

4Al - 4Mn ALLOY FREQUENCY DISTRIBUTION
ULTIMATE TENSILE STRENGTH AT ROOM TEMPERATURE
- BLADES, VENDOR A
- BLADES, VENDOR B
X̄ - 151,000 P.S.I.
σ 8,500 P.S.I.
VENDOR A X̄ - 148,000 P.S.I.
VENDOR A σ - 8,000 P.S.I.
SPECIFICATION LIMIT
NUMBER OF TESTS
120
100
80
60
40
20
0
120
130
140
150
160
170
ULTIMATE TENSILE STRENGTH - 1000 P.S.I.

Fig. 37


4Al - 4Mn ALLOY FREQUENCY DISTRIBUTION
0.2 PERCENT YIELD STRENGTH AT ROOM TEMPERATURE
- RINGS
- DISCS
$\overline{X}$ - 143,000 P.S.I.
σ - 5,000 P.S.I.
SPECIFICATION LIMIT
NUMBER OF TESTS
0
20
40
60
80
100
120
120
130
140
150
160
170
0.2 PERCENT YIELD STRENGTH - 1000 P.S.I.

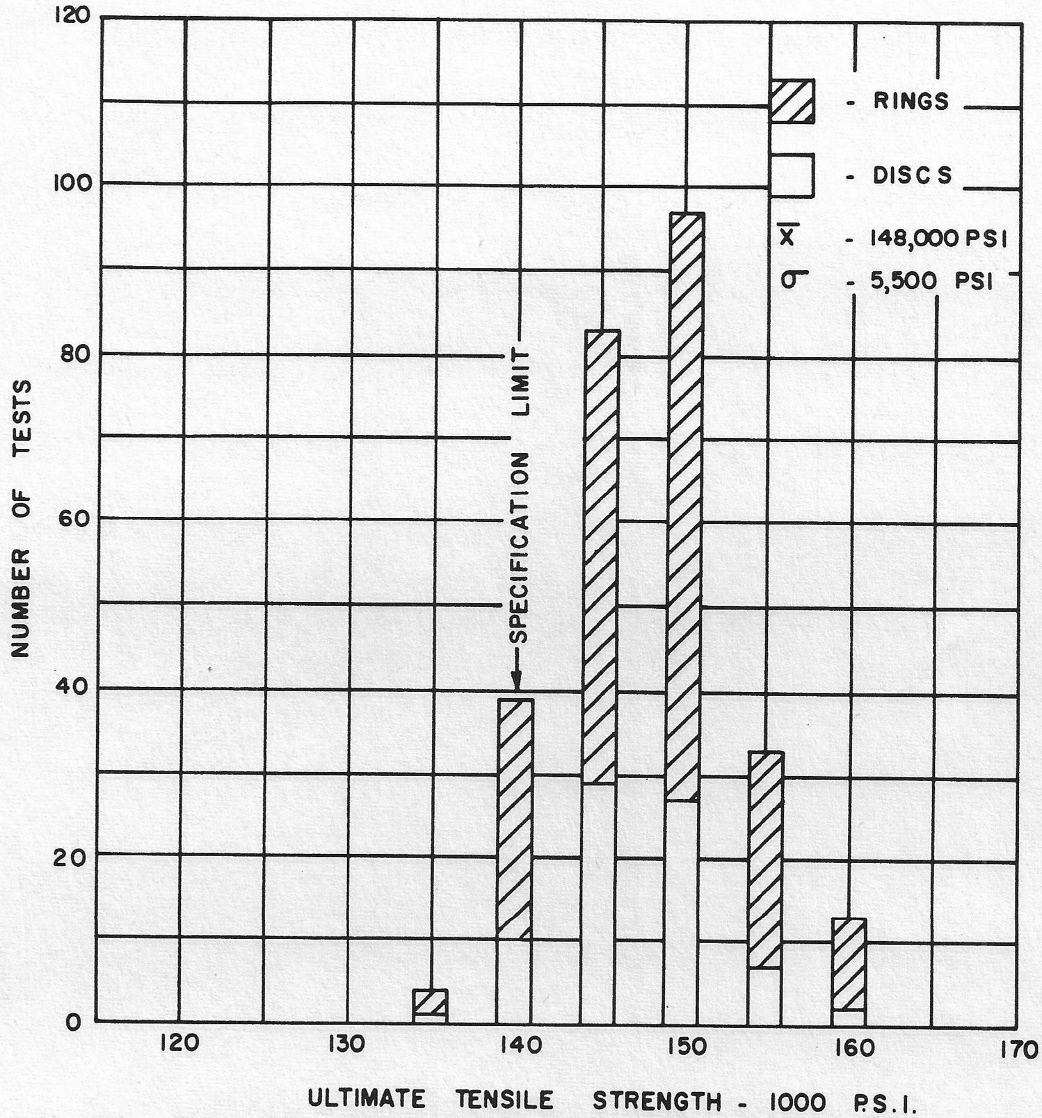

4 Al - 4 Mn ALLOY FREQUENCY DISTRIBUTION
ELONGATION AT ROOM TEMPERATURE
- RINGS
- DISCS
$\overline{X}$ - 18 PERCENT
σ - $2\frac{1}{2}$ PERCENT
SPECIFICATION LIMIT
NUMBER OF TESTS
0
20
40
60
80
100
120
6
8
10
12
14
16
18
20
22
24
26
ELONGATION - PERCENT

4Al - 4 Mn ALLOY FREQUENCY DISTRIBUTION
REDUCTION OF AREA AT ROOM TEMPERATURE



4Al - 4Mn ALLOY FREQUENCY DISTRIBUTION

ULTIMATE TENSILE STRENGTH AT ROOM TEMPERATURE



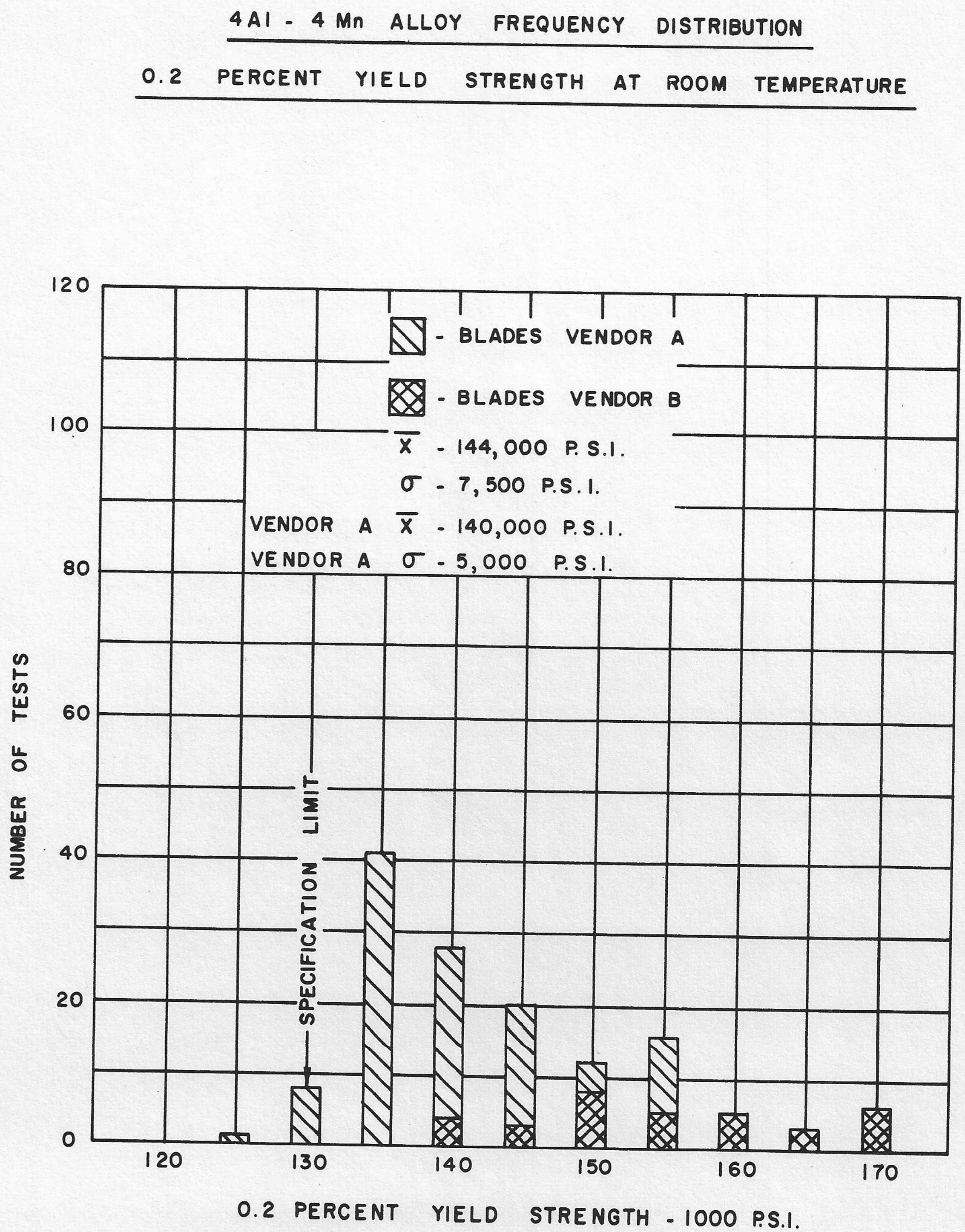
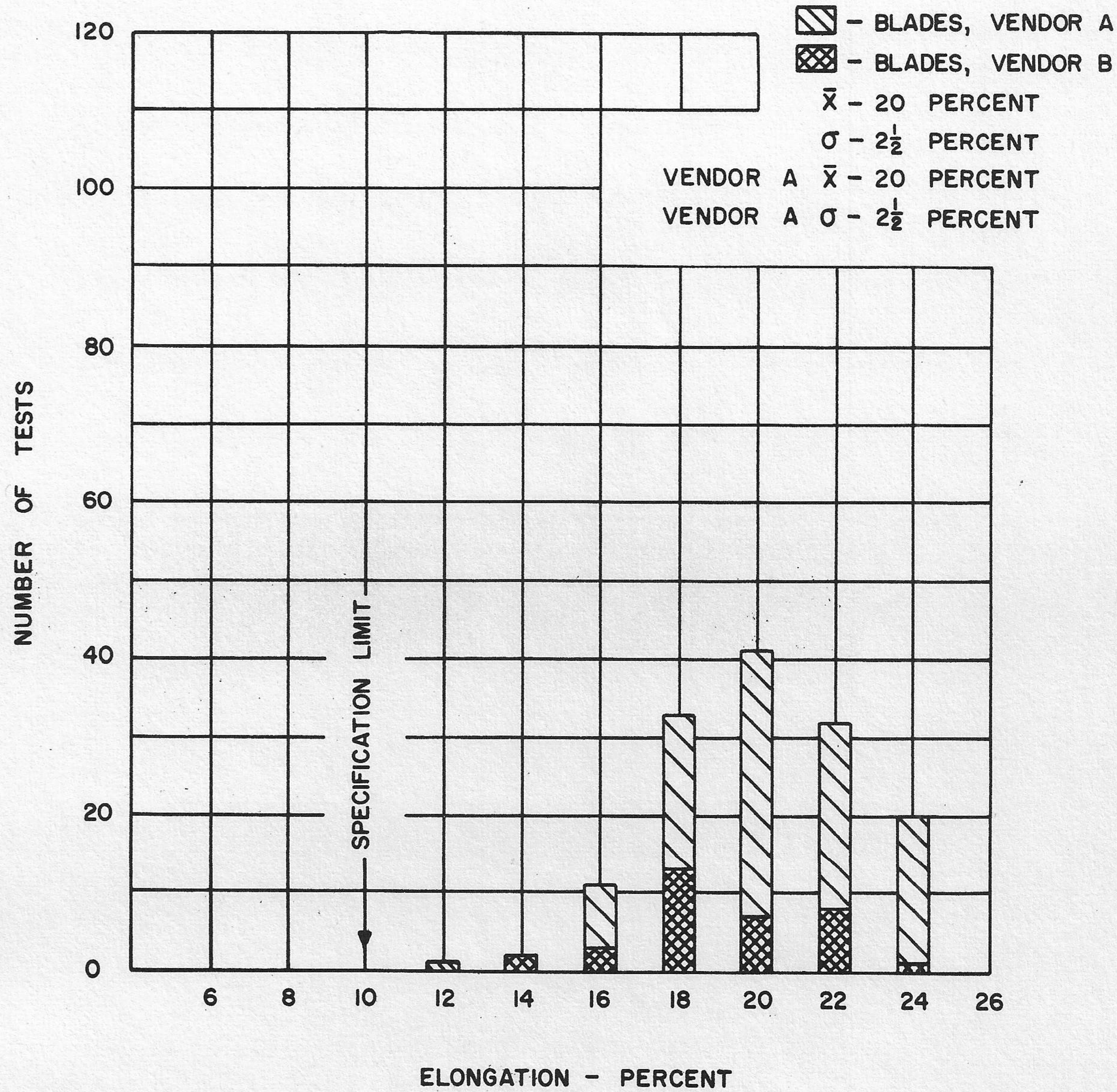

4Al - 4 Mn ALLOY FREQUENCY DISTRIBUTION
0.2 PERCENT YIELD STRENGTH AT ROOM TEMPERATURE
- BLADES VENDOR A
- BLADES VENDOR B
$\overline{X}$ - 144,000 P.S.I.
σ - 7,500 P.S.I.
VENDOR A $\overline{X}$ - 140,000 P.S.I.
VENDOR A σ - 5,000 P.S.I.
SPECIFICATION LIMIT
NUMBER OF TESTS
0
20
40
60
80
100
120
120
130
140
150
160
170
0.2 PERCENT YIELD STRENGTH - 1000 P.S.I.

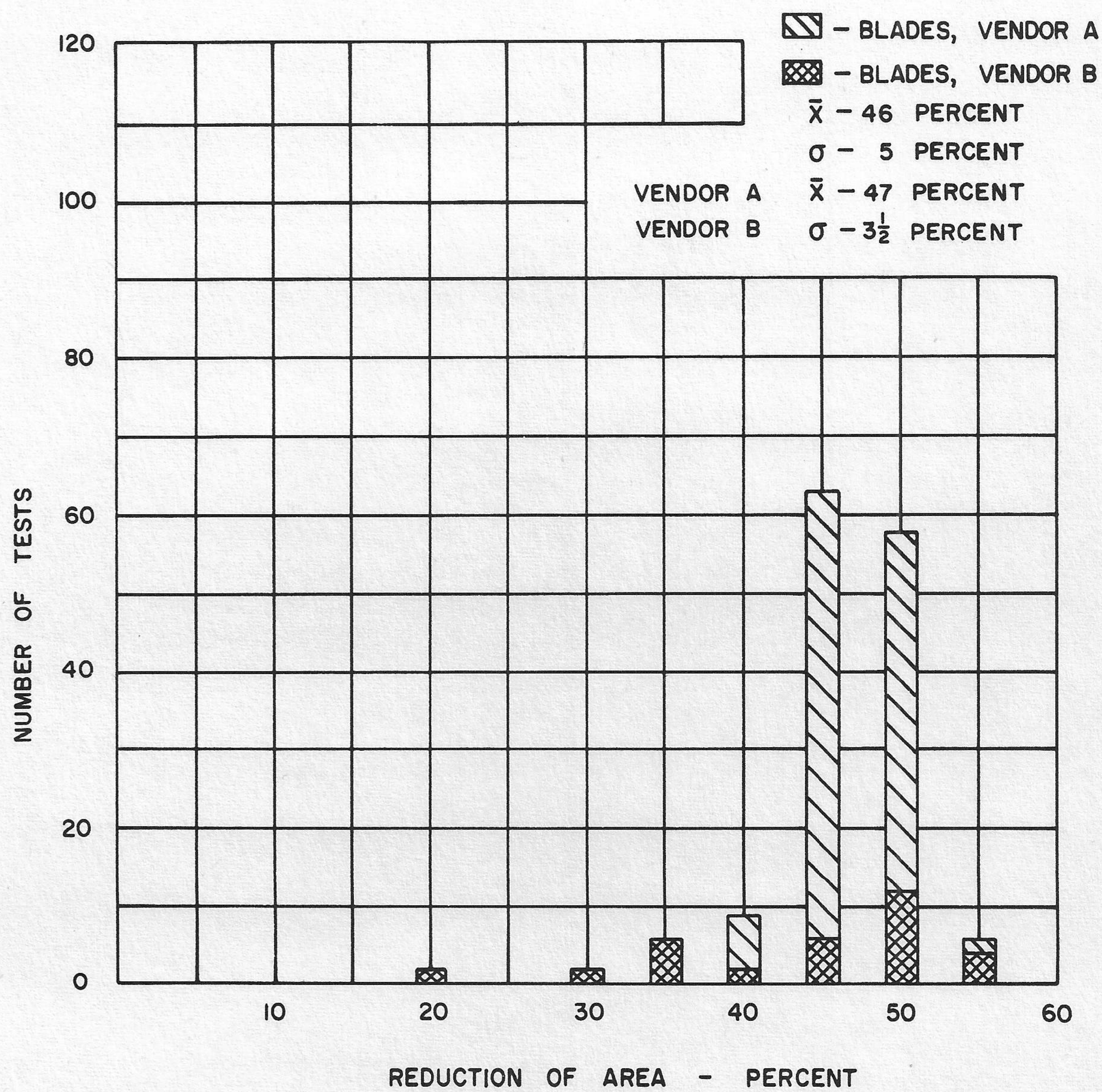
Fig. 42

4 Al - 4 Mn ALLOY FREQUENCY DISTRIBUTION

ELONGATION AT ROOM TEMPERATURE



4Al - 4 Mn ALLOY FREQUENCY DISTRIBUTION
REDUCTION OF AREA AT ROOM TEMPERATURE



difficult to assess, since as little as 0.05 percent oxygen (the margin of error in analysis technique) may account for 5,000 to 10,000 psi. Further, the influence of oxygen on tensile properties is lost at elevated temperatures. For these reasons the comparison of the different titanium alloys on the basis of room temperature properties is meaningless.

In general, the properties of the smaller blade forgings, due to the greater working they receive are better than those in heavy forgings. An interesting trend may be noticed between Vendor A and Vendor B. The properties of blade forgings from the latter seem to show a slight superiority, but insufficient test values are available to confirm this trend.

Figure 45 is a scatter band showing the relationship of elongation to reduction of area. The fractures in 4Al-4Mn alloy tensile test specimens, after deforming uniformily over the gauge length, sometimes displayed ovality or were otherwise non-symmetrical as we would expect from other metals. Hence good correlation between these two properties does not exist. This phenomenon is possibly due to the non-linear grain flow in the test specimen or the mixed hexagonal/cubic crystallographic lattice of titanium (the phenomenon has been noticed in magnesium, which also has an hexagonal lattice). In any event, elongation is the controlling measure of ductility with reduction in area being reported for information purposes.

Some interest has been shown in the relative merits of 6Al-4V alloy versus the 4Al-4Mn alloy. The 6Al-4V alloy in terms of ease of processing has much to recommend it e.g. relative freedom from freezing segregation during mélting, less sensitivity to deviations in heat treatment (untransformed beta), higher hydrogen tolerance, reportedly better forging flow characteristics etc. However, we are concerned here with mechanical properties.

Fatigue, particularly notched fatigue is of great interest to the design engineer. Titanium alloys have excellent fatigue properties and notch insensitivity in comparison with other metals. Interpretation of fatigue data, particularly with a limited number of tests may be perilous due to the inherent scatter in the material, and the problems inherent in specimen preparation and testing technique. While not fully confirmed, figure 47 does indicate that the two alloys are equivalent in smooth bar properties and that 6Al-4V may be superior in notched fatigue resistance. It should be noted that the notch fatigue characteristics of both alloys are superior to those of high strength steel. The curves are interesting for another reason for they show, like steel but unlike most non-ferrous alloys, a true endurance limit.

From the data shown, supported by tests at Orenda Engines Ltd.,

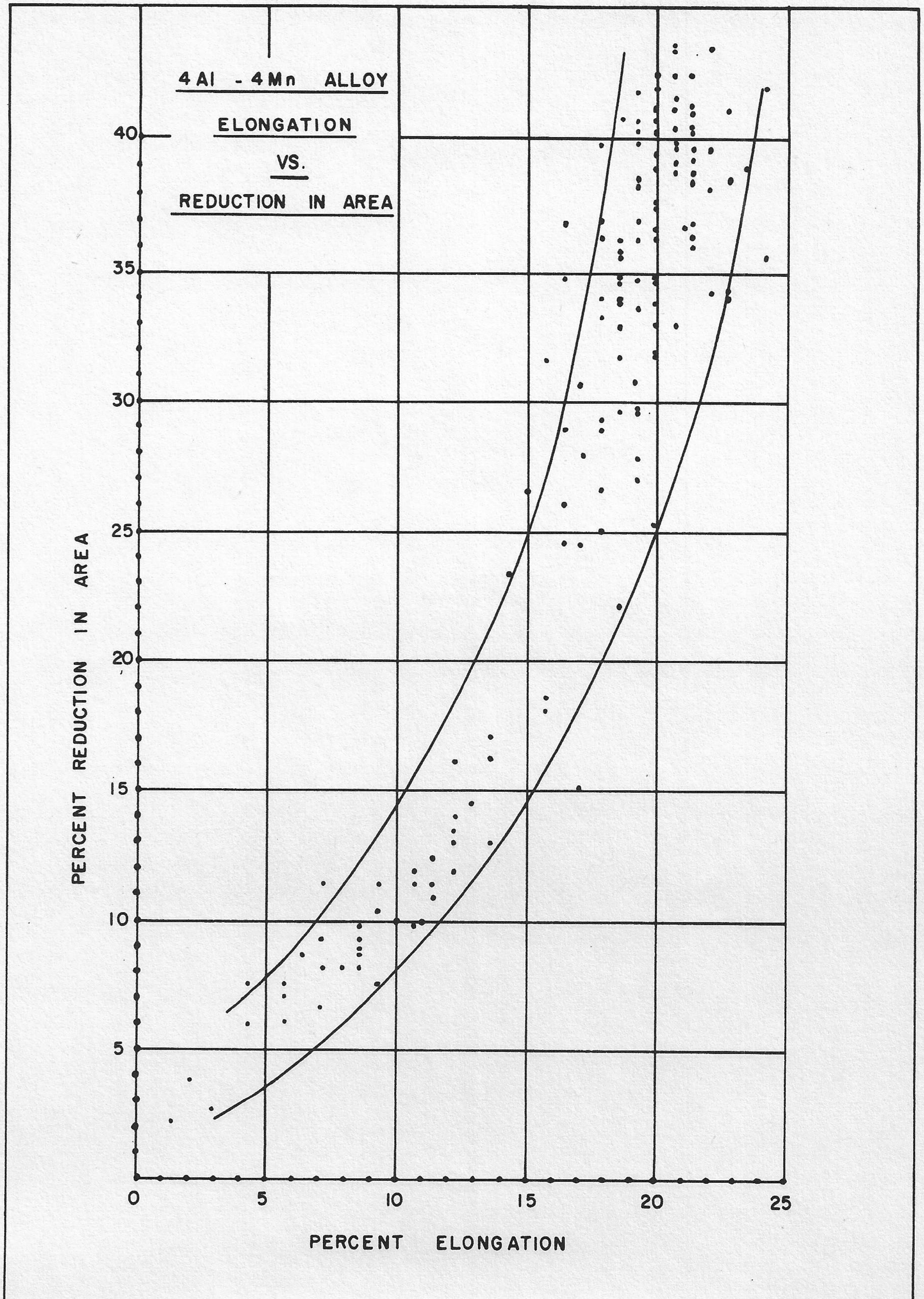

4 Al - 4 Mn ALLOY
ELONGATION
VS.
REDUCTION IN AREA
PERCENT REDUCTION IN AREA
40
35
30
25
20
15
10
5
0
5
10
15
20
25
PERCENT ELONGATION

Fig. 45

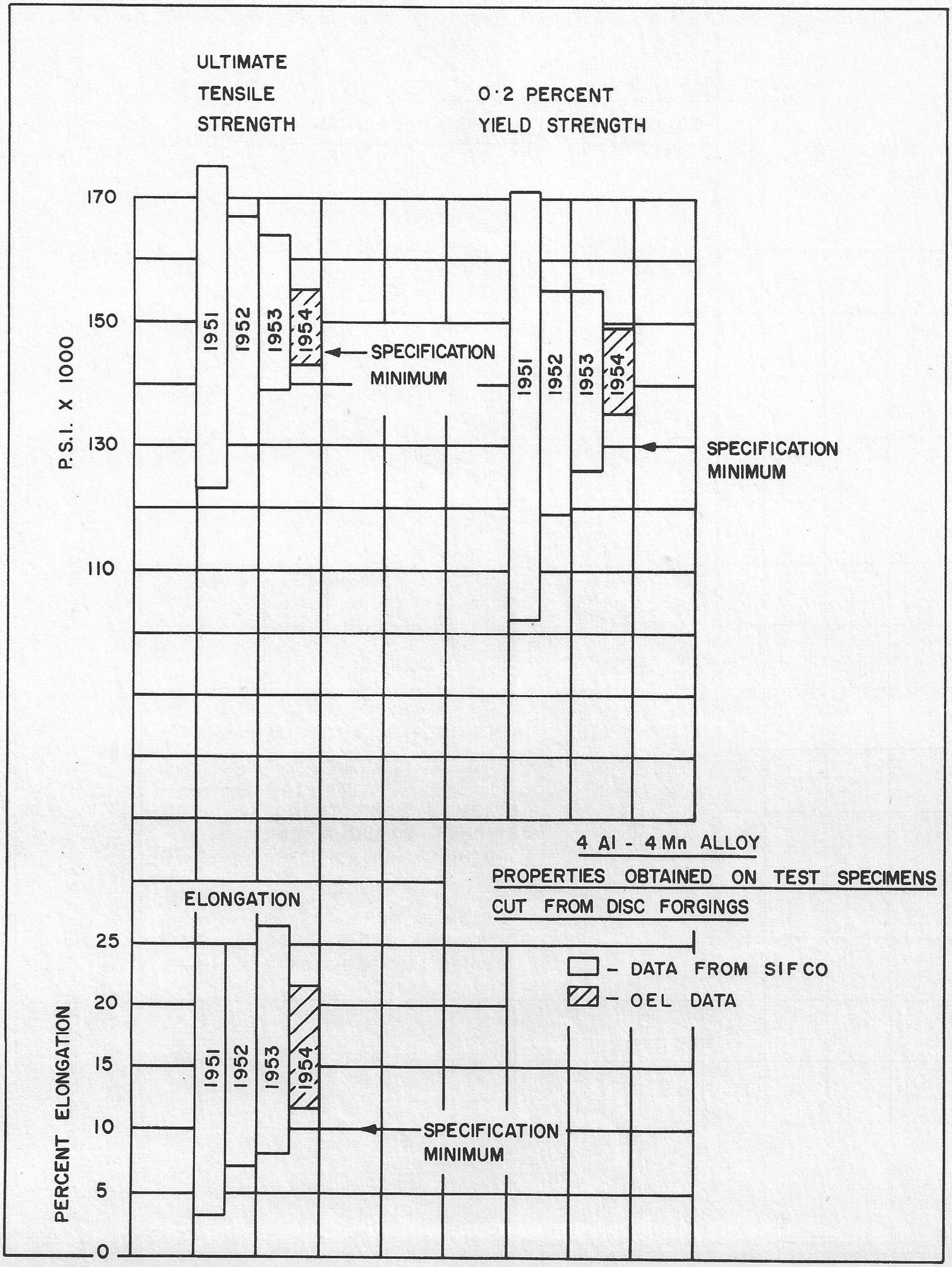

ULTIMATE TENSILE STRENGTH
0·2 PERCENT YIELD STRENGTH
P.S.I. X 1000
170
150
130
110
1951
1952
1953
1954
SPECIFICATION MINIMUM
SPECIFICATION MINIMUM
4 Al - 4 Mn ALLOY
PROPERTIES OBTAINED ON TEST SPECIMENS CUT FROM DISC FORGINGS
ELONGATION
PERCENT ELONGATION
25
20
15
10
5
0
- DATA FROM SIFCO
- OEL DATA
SPECIFICATION MINIMUM

Fig. 46

CONFIDENTIAL

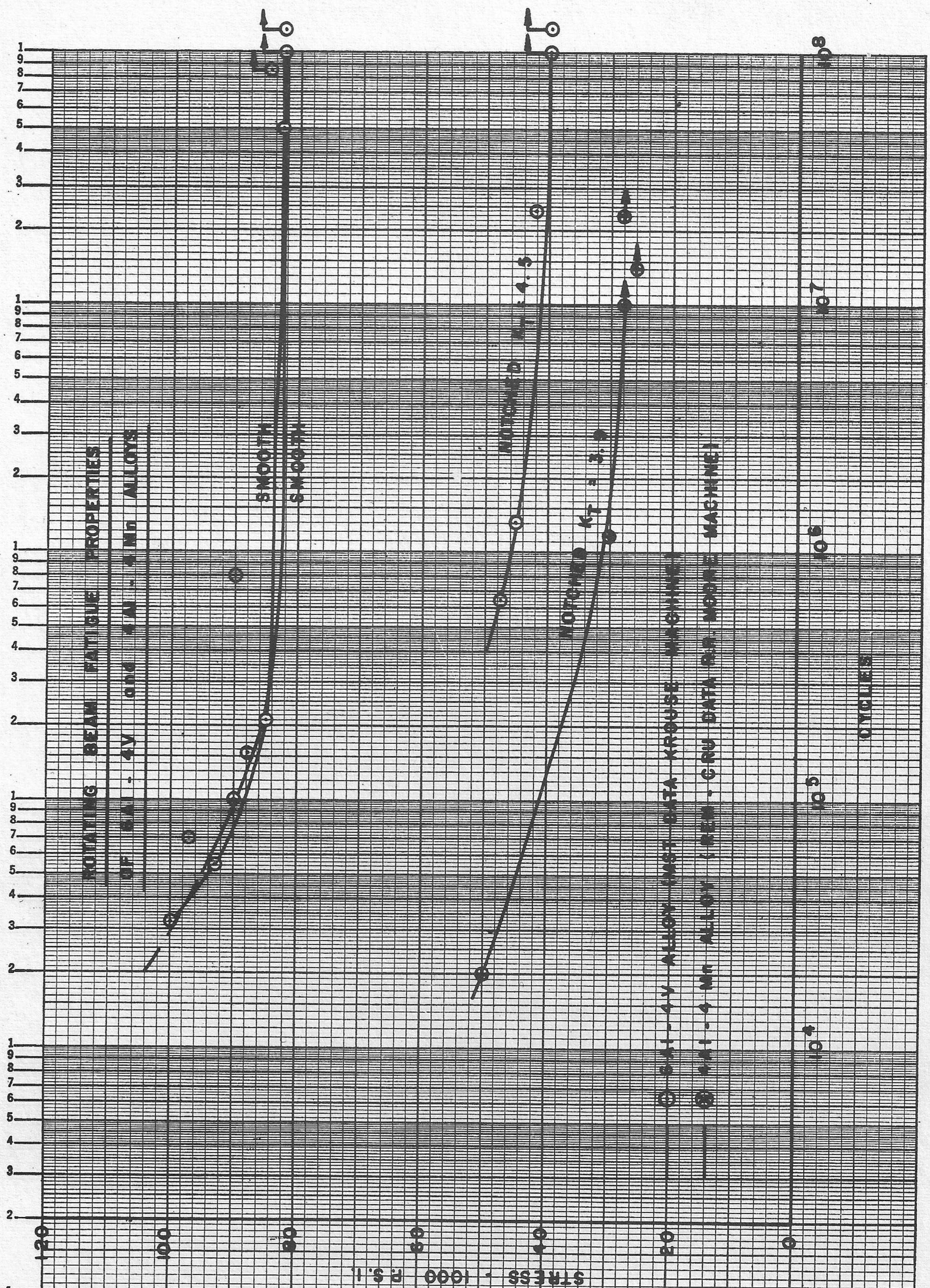


Fig. 47

the fatigue characteristics of the two alloys may be presented as follows:-

Alloy	Endurance Ratio	K_t	K_f	"q"
4Al-4Mn	0.50 to 0.60	3.9	2.78	0.61
6Al-4V	0.50 to 0.60	4.5	2.50	0.40

where Endurance Ratio = the ratio of $\frac{\text{smooth bar endurance limit}}{\text{ultimate tensile strength}}$

K_t = theoretical stress concentration factor of the notch

K_f = the fatigue strength reduction factor due to the notch

"q" = the notch sensitivity index or $\frac{K_f - 1}{K_t - 1}$

The influence of surface finish is pertinent to any consideration of fatigue. The following data was developed in our laboratory to assess this factor.

TABLE IX

The Effect of Various Surface Treatments on the Endurance Limit of 4Al-4Mn Alloy

Group	Surface Preparation	Endurance Limit (PSI)
A	fine hand polishing	± 100,000
B	shot peened to 0.012A2 intensity	± 91,000
C	vapour blasted	± 97,000
D	grit blast #60 grit	± 83,000
E	80 grit wheel dressed with "N" cpd and Formax F122	± 78,000
F	same as E, then vapour blasted	± 91,000
G	same as E, but abused by overheating with dry cpd	<± 85,000
H	electropolished	>± 80,000

Although this exercise was of limited scope and therefore questionable from the statistical viewpoint, the trends observed lead to certain recommended practices in processing blade forgings or other parts critical with respect to fatigue:-

(a) Vapour blasting (liquid honing) is particularly effective in improving fatigue resistance (groups E and F).

(b) Vapour blasting of a machined surface approaches hand polishing in effectiveness (groups A and C).

(c) The fatigue resistance is proportionally reduced by the severity of the finishing operations (groups A and E) as might be expected. Fine machining is therefore desirable.

(d) Sand or grit blasted surfaces on finished forgings received from the Vendors are not the optimum condition for service (group D). Hand polishing or vapour blasting are much to be desired as the final finishing operation.

(e) Electropolishing offers little assurance that the fatigue properties will not be impaired.

(f) Shot peening does not offer much hope of improving the fatigue properties of titanium. However, this data is based on extended life (15 X 10^6 cycles); for a shorter number of cycles at stresses above the endurance limit, some improvement is realized by shot peening.

Other investigators have reported on the effects of grinding which, aside from any consideration of burning or localized overheating (an ever present potential problem), can lead to severe reduction (30 percent plus) in fatigue properties due to residual tensile surface stress.

Little or no data exists on fatigue properties at elevated temperatures, i.e. the area in which the engineer is most concerned, although the data that has been reported, has shown no pronounced drop off in the 4Al-4Mn Alloy until 800 degrees F is reached. Assuming a constant endurance ratio, the tensile properties become of interest (figure 48).

It can be seen that at temperatures greater than 325 degrees C, 6Al-4V has equivalent or superior properties to 4Al-4Mn. Below this temperature especially in considering minimum guaranteed levels, the 4Al-4Mn shows to advantage, (10,000 psi at 20 degrees C). The higher properties of the 4Al-4Mn at lower temperatures may be attributed to the slightly higher oxygen content (a nominal 0.20 percent versus 0.15 percent in 6Al-4V the strengthening effect of which could easily account for 5,000 to 10,000 psi), the greater amount of beta in this alloy or the relative strengthening effect of manganese versus vanadium of the beta phase. The fact that this strength advantage is lost at higher temperatures suggests that it is due entirely to the slightly higher oxygen level in the 4Al-4Mn alloy. It is noted that oxygen affects the notch toughness and residual ductility under creep strain conditions of alpha-beta alloys.

Subject to further experience, it may be possible to raise the

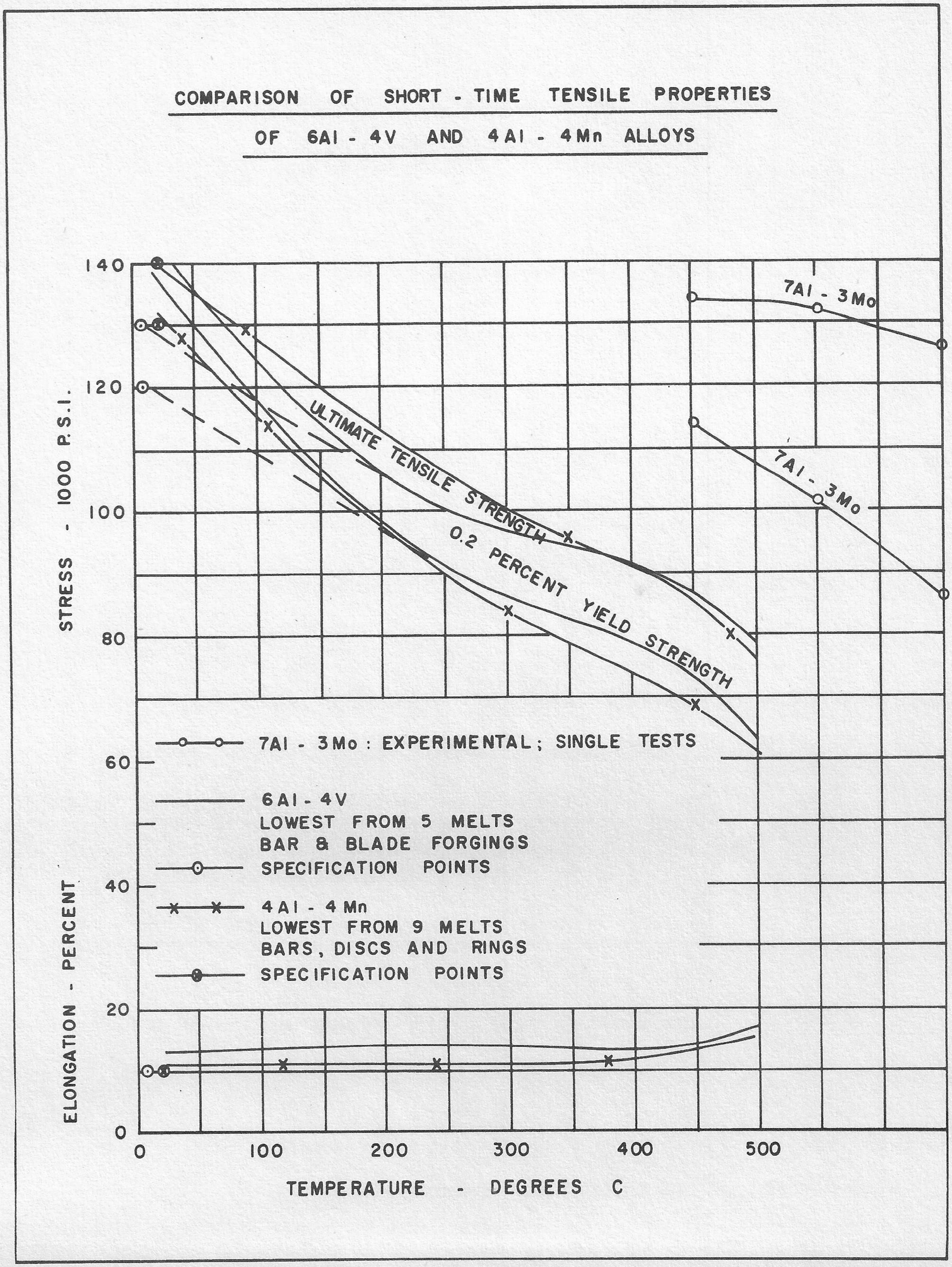

COMPARISON OF SHORT - TIME TENSILE PROPERTIES
OF 6Al - 4V AND 4Al - 4Mn ALLOYS
STRESS - 1000 P.S.I.
ELONGATION - PERCENT
TEMPERATURE - DEGREES C
140
120
100
80
60
40
20
0
100
200
300
400
500
7Al - 3Mo
7Al - 3Mo
ULTIMATE TENSILE STRENGTH
0.2 PERCENT YIELD STRENGTH
7Al - 3Mo : EXPERIMENTAL; SINGLE TESTS
6Al - 4V
LOWEST FROM 5 MELTS
BAR & BLADE FORGINGS
SPECIFICATION POINTS
4Al - 4Mn
LOWEST FROM 9 MELTS
BARS, DISCS AND RINGS
SPECIFICATION POINTS

Fig. 48

specification minimum limit for the 6Al-4V alloy blade forgings to an equivalent level with the 4Al-4Mn alloy, based on the general improvement one gets with smaller, more heavily worked sections. This would give a more favourable representation for 6Al-4V but would not alter the basic difference in nominal properties at room temperature. The nominal yield strength and creep properties (though not the low temperature minimum limits) are seen to favour the 6Al-4V alloy with the temperature limit of about 315 degrees C being the point above which 6Al-4V has a clear advantage with no qualifications (figure 49). As previously mentioned, comparison of the two alloys at room temperature is somewhat meaningless in view of the influence of oxygen content.

Among the new alloys of potential interest to the engine manufacturer are those of the Al-Sn series (e.g. I.C.I. Alloy 371 12 percent Sn 2½ percent Al) and the Al-Mo series (e.g. 7Al-3Mo, 6Al-4Mo, 6Al-3Mo-V). Although some limited laboratory experimental data for one of these, 6Al-4Mo is shown in figure 48, they have yet to be evaluated in terms of heat treatment, forgeability, stability and commercial feasibility in full scale ingots. This work is currently in progress. The objective is a satisfactory titanium alloy for use at temperatures up to 550 degrees C.

6.0 SERVICE EXPERIENCE

The following photographs have been included to illustrate various metallurgical aspects in the application of titanium alloys in aircraft power plants. While quality control must ever be a constant problem if titanium alloys are to be brought to a high degree of uniformity and reliability as an engineering material, the final behaviour in service is the criterion of acceptability of this material.

Figure 50 illustrates an important characteristic in the blade failure and in engine X101/4, quite aside from hydrogen sensitivity problems previously covered. It is apparent that fretting on the root face created the nucleus for a fatigue crack.

The anomalous frictional characteristics of titanium have long been recognized. Titanium has a low coefficient of friction (0.47; steel on steel is 0.53) but a pronounced tendency towards severe galling and metal transfer (pressure welding) when the atmospheric oxide film (which forms initially on heating) breaks down, which is quickly. Conventional lubricants such as palmitic acid in cetane, graphite, molybdenum disulphide etc. are generally inadequate in preventing galling since no effective absorption of the lubricant molecules takes place on the titanium surface.

CONFIDENTIAL

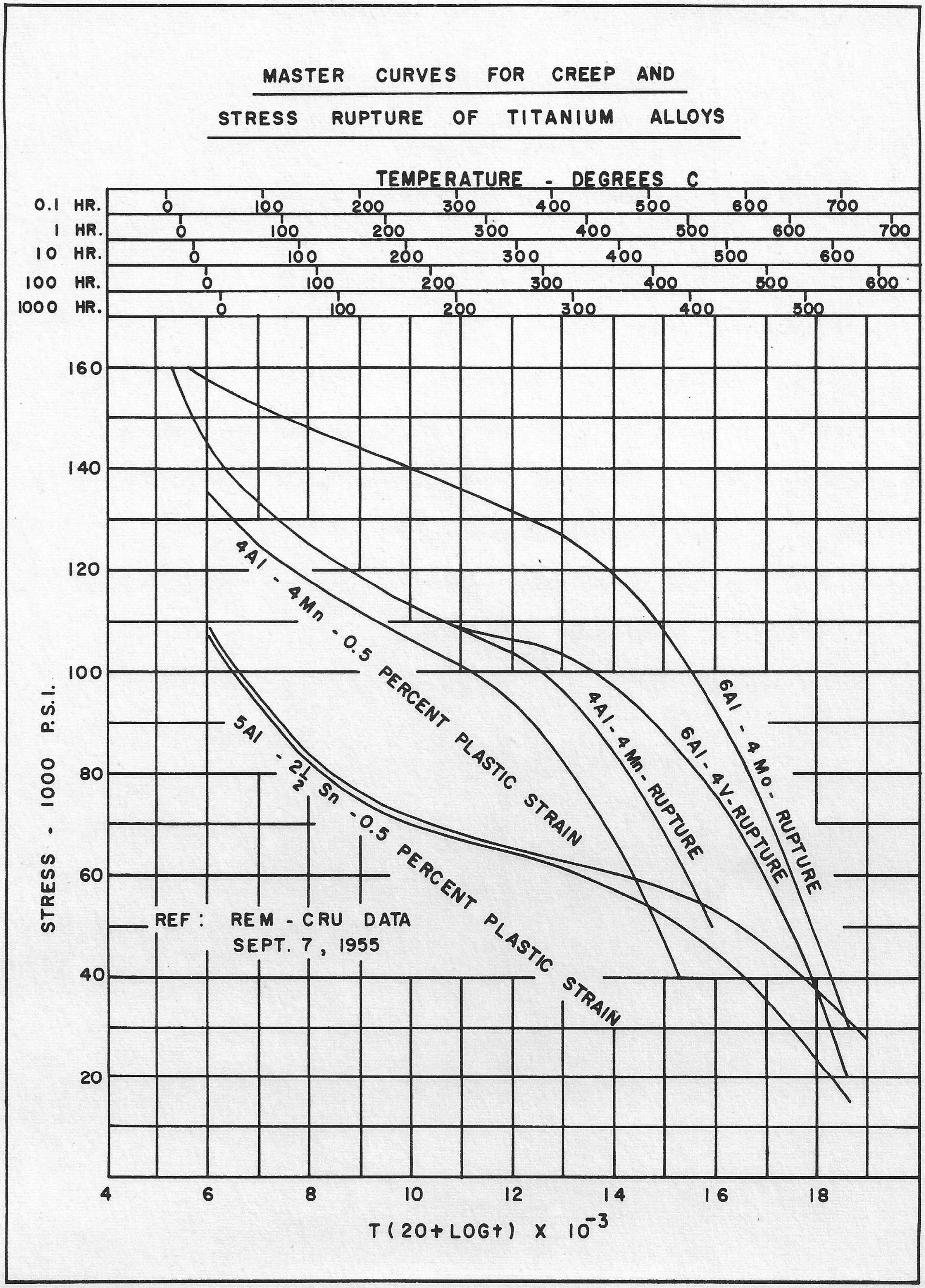


Fig. 49

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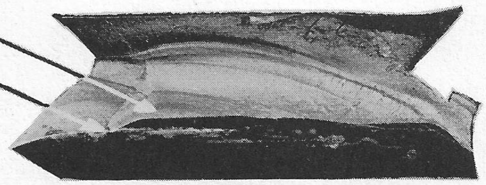


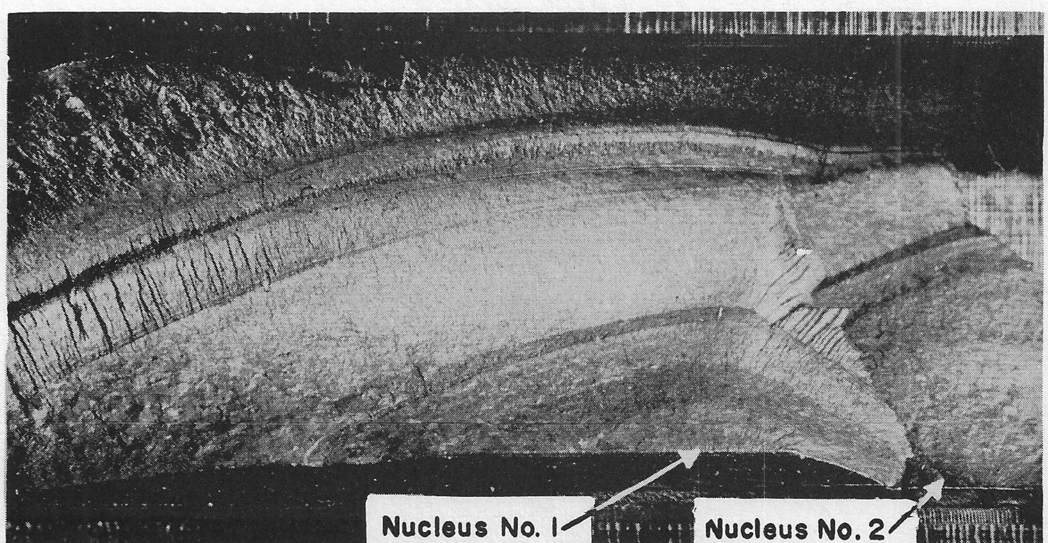
Blade No. 17 Code CZL Part No. 80004

Engine X 101/4

X 1.5

BLADE ROOT FATIGUE - FAILURE, X 101/4

35 hours service - high hydrogen Fe-Cr-Mo Alloy — note fretted area and location of Nucleus No. 2 with respect to fret work.



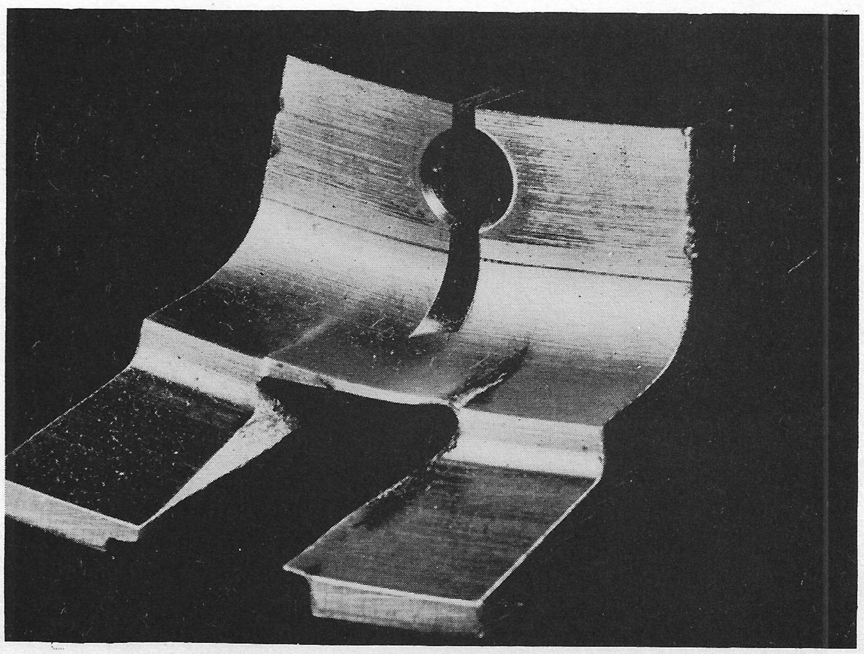
X 5

Consequently, the occurrence of fretting on the blade roots, leading to fatigue failures is a serious problem. Laboratory work with or without vibration rig tests on various anti-fretting measures for this application have narrowed the interest down to solid lubricants where epoxy resin type is used as a vehicle for graphite or molybdenum disulphide. This type of lubricant not only has low shear strength and hence low friction characteristics but more important, it is supported laterally in service and hence resists displacement. The pretreatment of the base material is important. Shot peening or grit blasting may be useful as a keying base; chemical conversion undercoats of the phosphate/fluoride type while less effective than shot peening for mechanical keying are considered to have good anti-fretting and temperature resistant characteristics.

Silver plating of blade roots has received a high merit rating on blade vibration rig tests. This process uses a nickel strike with grit blasting and chemical cleaning. The bond so formed is of a mechanical/chemical nature which has not proven 100 percent reliable on scaling up from the laboratory to shop operations. Erratic bonding results are obtained on any one batch of blades. Various types of precleaning treatments such as hydrofluoric/ethylene/glycol, anodic etching and diffusion of the nickel by heat treating are currently under investigation. Plating from fused salts is also a promising but untried avenue. In view of these chemical problems i.e. the maintenance of an initial oxide free surface on titanium for plating, it must be said at this date, that silver plating must be excluded as a reliable repetitive shop operation.

It is essential in a highly stressed component such as a compressor disc, to have sufficient ductility (as well as high yield strength and creep resistance) to permit local plastic yielding and the relief of peak stresses due to design stress raisers. Figure 51, a photograph of the rim of the failed disc from engine X101/11 shows the high degree of yielding which occurred in this area without evidence of cracking.

The engine designer as part of his assessment of a material is obliged to consider the toughness (impact resistance) of a proposed material. This is viewed not only from the aspect of foreign object damage, but also from the viewpoint of secondary damage following the initial failure of a component. The photograph, figure 52, of three blades removed from a failed compressor shows that 4Al-4Mn alloy has satisfactory ductility and impact resistance for such applications. This is further demonstrated in the failed disc shown as figure 53.



Slot No. 27 Part No. 80004

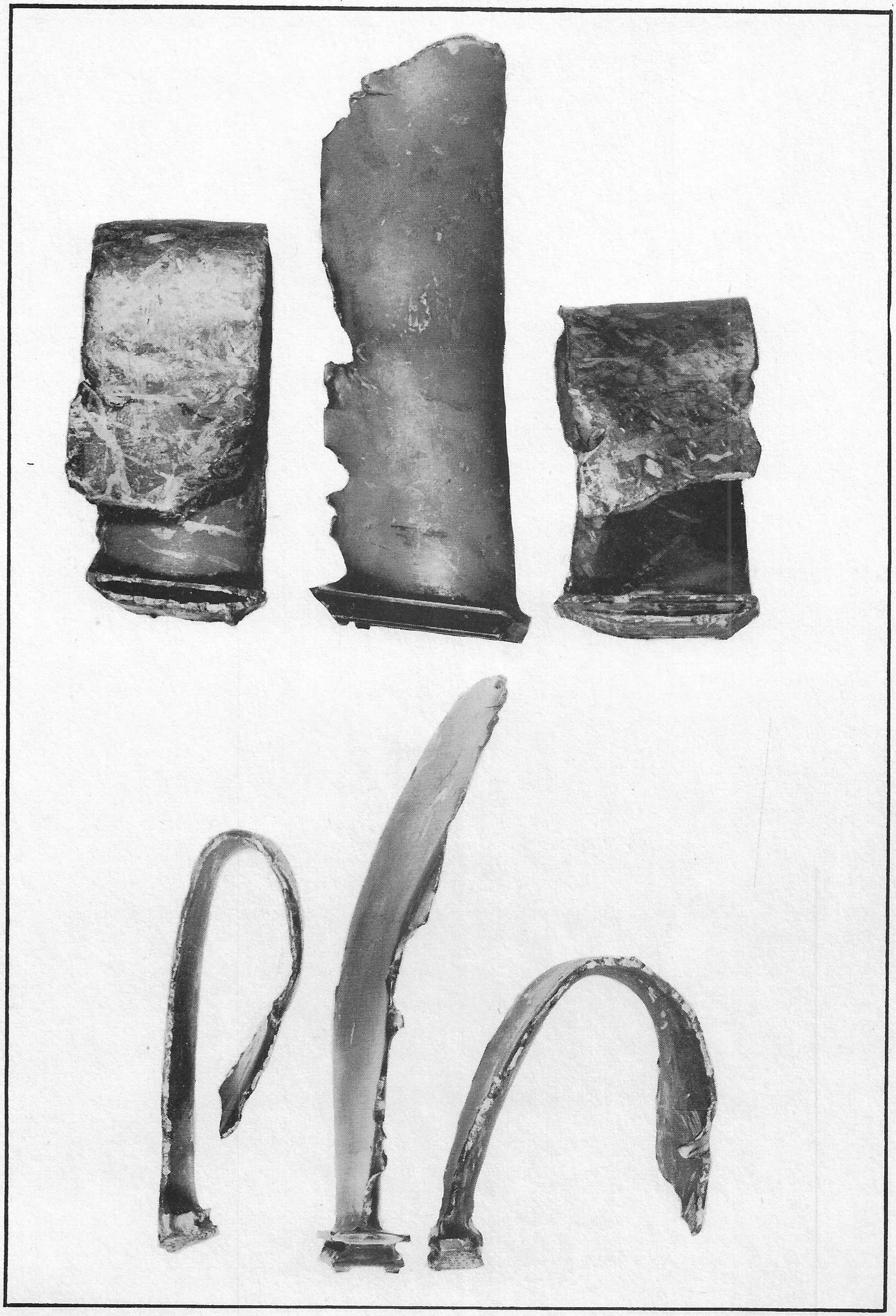
Engine X 101/4

X 1.75

COMPRESSOR DISC

Showing dented surface at slot.

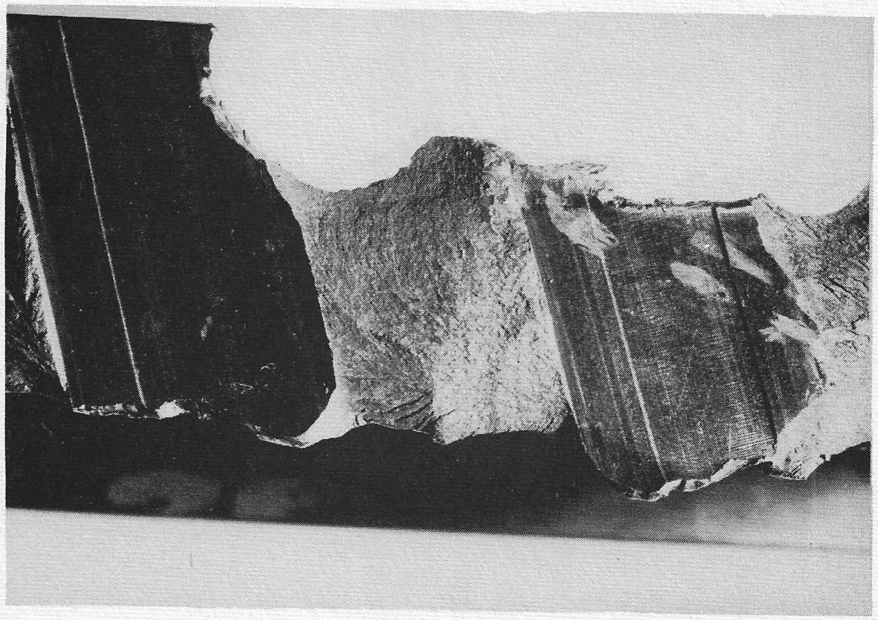
Fig. 51



BLADE DAMAGE FOLLOWING ENGINE FAILURE

Note high degree of ductility and resistance to impact.

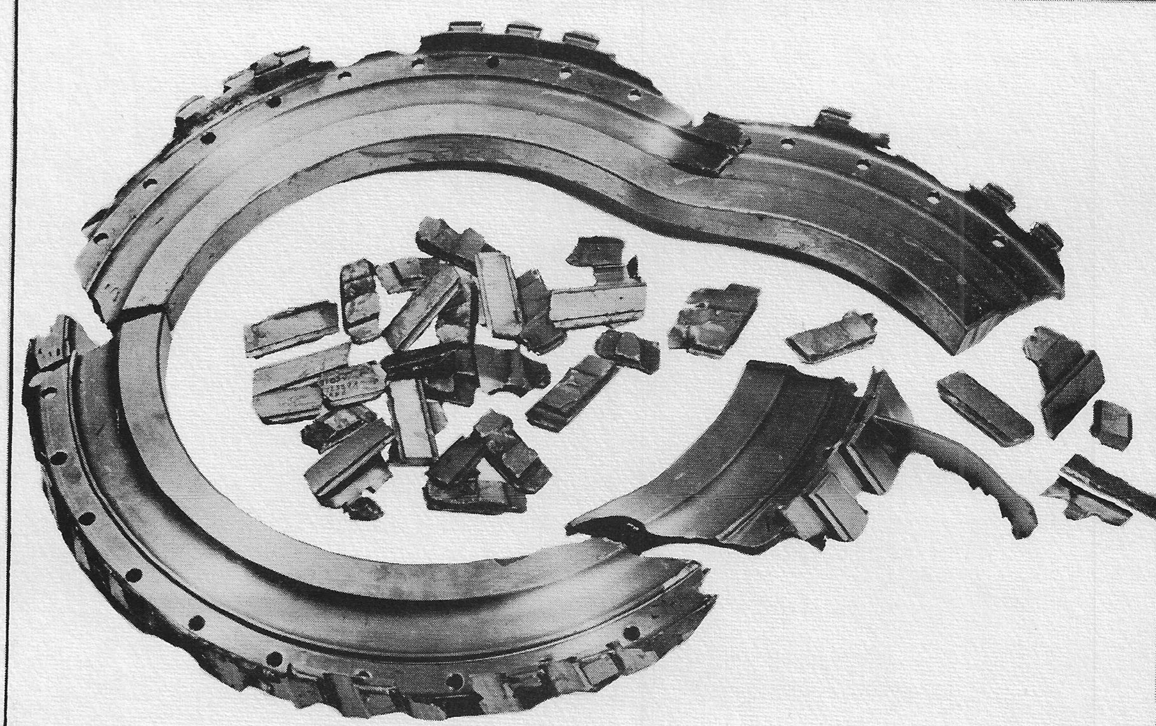
Fig. 52



X 1 1/2

Fatigue fracture at base of inter blade land from X 101/11 ——

111 hours service 4Al-4Mn Alloy



4th. Stage Compressor Disc from X101/11

Note ductility of hub - secondary damage

Fig. 53

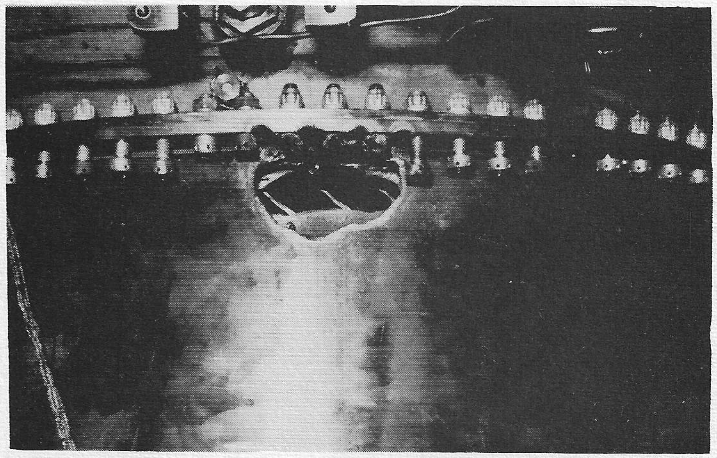
The importance of adequate notch fatigue resistance in aircraft engine applications is illustrated by figure 53, the area of initial failure in engine X101/11 after 111 hours running time. This is a classical textbook example of fatigue showing clearly defined progression lines (beach marks). Hence our concern in laboring the fatigue characteristics of titanium alloys in the preceding paragraph.

Titanium spark testing is distinctive for the intense spark stream generated. Although titanium has excellent flame resistance, and once lighted, is self-extinguishing in the massive form, it will display the usual burning or even explosive qualities (as dust or flour) of a finely divided material.

The pyrophorric qualities and low thermal conductivity characteristics of titanium led directly to burning experienced in engine X101/5 when a failed 10th stage blade became wedged at the spacer ring. Because of the persistence of the source of heat, the fire continued to burn until a hole was well established in the casing (figure 54). On shut down, the fire extinguished itself but, the potential problems of an engine within an airframe where a blow torch effect follows a blade failure may be appreciated. As serious as this problem was, we were hardly prepared for the ultimate catastrophic result (figure 55) which occurred when several components within the compressor broke up.

The secondary fire hazard of an all titanium compressor is only too apparent. It is doubtful if any form of surface treatment (sprayed ceramic coatings, etc.) could effectively act as a thermal barrier in containing the intense heat of a titanium fire. Since the physical properties are unalterable, any credit for eliminating or containing this hazard will have to lie with the mechanical ingenuity of the designer. Strangely enough, to-date there has been no reported parallel experience with U.S. engine manufacturers experimenting with compressor configurations in titanium, one of which is reported to be in production.

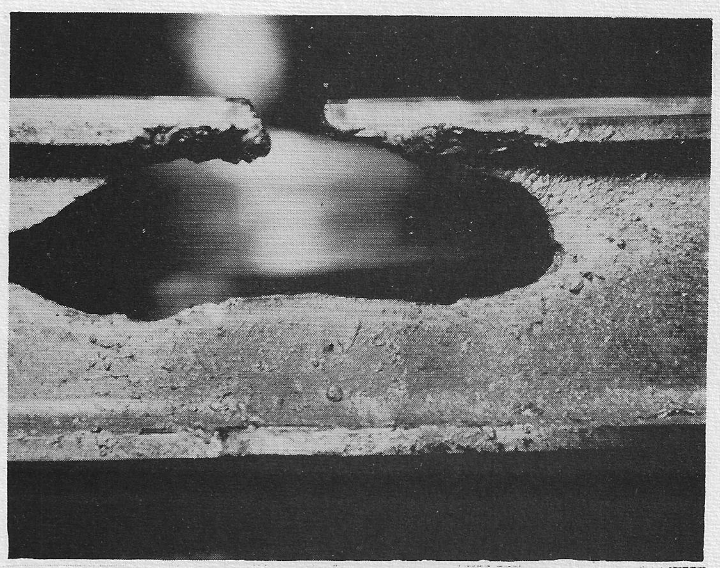
To reject the possibility of a solution to the fire hazard problem at this time is to deny the many desirable advantages that are realized by the use of titanium in aircraft gas turbines. Indeed, at the time of writing, with limited experience on relatively few schemes, the fire hazard problem seems to be a past experience.



Photographs Showing Fire Damage From X 101 /5 —

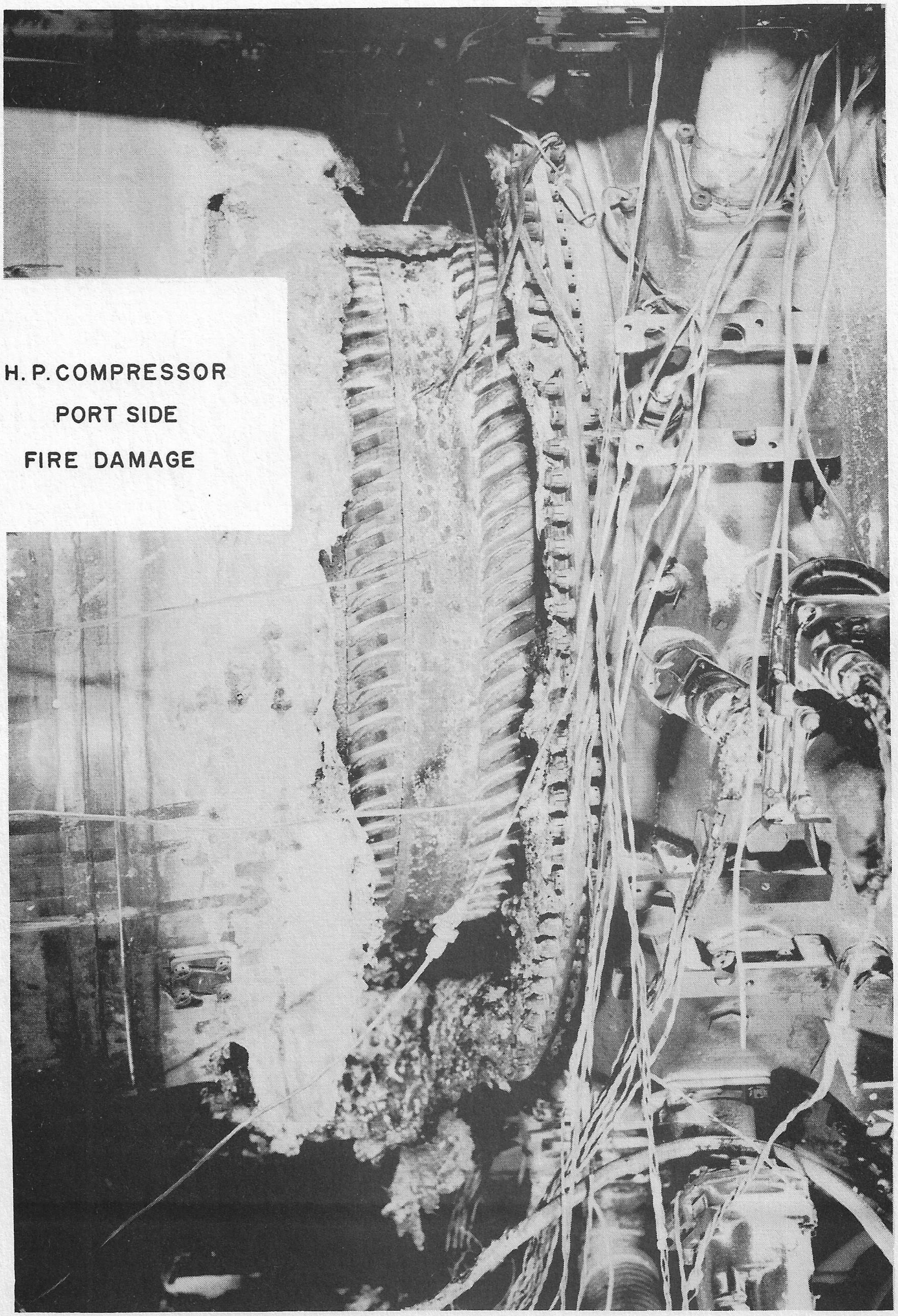
High pressure casing (C.P.Titanium)
above

Stator spacer ring (Fe-Cr-Mo Alloy)
below



Burning followed the failure of 10th stage blade containing a forging defect after 22 hours service.

Fig. 54



Extensive Fire Damage In X 101/12
4th. Stage Disc Failure After 127 Hours Running ——

Fig. 55